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RAY T. BAYLESS, *Editor*

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THE SCIENTIFIC METHOD IN METALLURGY

By S. L. HOYT

PROFESSOR Edward De Mille Campbell, whom we honor each year at this hour, was pre-eminently a scientist. To him the scientific method was sound and logical and it was natural for him to use it rather than the method of reasoning things out or even the more practical empirical method. In laying out his work, in his experimentation, in analyzing his results, and in speculating on the meaning of it all, he was the true scientist. Is it not appropriate then, in this tribute to his memory, to examine our chosen field in this same spirit; to trace its genealogy, its growth and evolution, its present status, and probable future trends? In doing so we need not be chagrined when we uncover shortcomings, while we may note with pardonable pride that metallurgy has a fine family tree with healthy roots in fertile soil, a sturdy trunk, and branches which reach outwards and upwards in all directions. It is for these reasons that I choose as the subject of the 15th Campbell Memorial Lecture—"The Scientific Method in Metallurgy".

The justification of this choice is based on the subject matter of metallurgy which now falls by common consent into the class of a science. In other words, metallurgy represents a co-ordinated branch of knowledge which permits ordering, classification, and interpretation according to recognized laws. This has not always been so and here we need but recall the alchemistic forerunner of metallurgy and the typical empirical methods of early metallurgical practice. Fortunately, though naturally enough, metallurgy has now come of age and I shall attempt to show that that splendid consummation was effected at the turn of the century as the culmination of many years of preparation both in metallurgy proper and in the sciences that are basic to it.

Mathematics is the most basic and rigorous of all the sciences, though curiously enough it did not produce the scientific method.

This is the Fifteenth Edward DeMille Campbell Memorial Lecture, presented by S. L. Hoyt, technical adviser, Battelle Memorial Institute, Columbus, Ohio. The lecture was presented at the Twenty-Second Annual Convention of the Society, held in Cleveland, October 21 to 25, 1940.

It was also the first of the sciences and the geometry developed by the Greeks several centuries before Christ is still taught in our schools. The less rigorous, and therefore more speculative, natural sciences made a relatively late start, while the highly speculative sciences of human behavior are only now being developed. Metallurgy, we may note, has been empirical and speculative and it is not surprising that it has only recently gone through the process of being synthesized into a science. Apparently the abstract, though rigorous, mathematical sciences submit more readily to pure reasoning and are more easily encompassed than are those sciences which involve observation or experimentation, and ordering according to basic laws. In this sense metallurgy has followed the historical pattern.

It is customary in such discourses to start with the Greeks and to allude to their contributions in science, literature, philosophy, politics, the arts, etc. Restricting ourselves to the sciences it is clear that their only lasting contribution was in the field of mathematics. There their remarkable ability, not to say facility, for logical reasoning bore fruit; starting with axioms and hypotheses, deductive reasoning led them to correct conclusions. So potent was this reasoning facility that they overlooked, we may surmise, the necessity of testing and proving their premises; on this account there is a lot on the negative side to be said of the Greek contribution to the natural sciences. There they set up the assumption in a completely arbitrary way that the material world was composed of earth, water, air, and fire. This idea has since then been dignified by some misguided souls with the appellation of an "atomic theory". The properties of matter were held to be due to something quite aside from its substance, though here their ideas were equally futile. A heavy body was "matter-stuff" with a large amount of "heaviness" and a buoyant body was the same matter-stuff endowed with a large amount of "lightness". Heaviness or lightness was an attribute, or the "essence", of those bodies; and that explained the situation. Entirely ignoring the implications of their assumptions and unaware of their difficulties, the Greeks apparently relied solely on their ability to make sound assumptions and on the facile technique of their deductive reasoning.

My discussion here is entirely inadequate, I realize, but essentially this philosophy of the natural sciences was developed and rationalized by Aristotle and history shows that it dominated the

thought of the civilized world for over 1500 years. No less a person than Goethe has said that Aristotle saw nature better than any modern, but was too quick with his opinions. The moral here, and I am sincerely of the opinion that an important lesson is to be had from this experience, is that a system can be established and accepted, and yet be devoid of useful subject matter. However, on wondering at this major fiasco, we might also speculate on the change which civilization probably would have taken had the Greeks been inspired to use the experimental method, instead of relying solely on their ability to "reason things out".

We now jump 1500 years of barren scientific history, over which the false philosophy of Aristotle and the mystic teachings of alchemy held full sway. The fervent plea of Roger Bacon for experimentation went unheeded and it was not until the time of his equally famous namesake, Sir Francis Bacon, that modern procedure was first used. This brings us to the Renaissance and the work of Galileo who has been called the first truly modern man of science. In founding mechanics or physics he also introduced the scientific method, and thereby broke the spell of Aristotle. Galileo held that the natural sciences must be studied by observation and controlled experimentation, and that both assumptions and conclusions are to be subjected to experimental verification. Thus history records the famous test case at Pisa where Galileo is supposed to have shown that two bodies, though differing in size, fell to earth at the same velocity. Aristotle had taught that the larger one, with more "heaviness", would fall faster, but he never tried the experiment. This theatrical demonstration dramatized the difference between the two schools, and taught, for all time, that experiment is to be substituted for assumption.

This was but one example of the schism that formed at that time between the old and the new in scientific thought. We may also recall the teachings of alchemy that metals are transmutable, the idea coming from peculiar assumptions as to the constitution of matter, or the teachings of astrology regarding the effects of heavenly bodies on the destinies of man on earth, both of which are traceable to the philosophy of Aristotle. My object in mentioning these points is not to comment on their absurdity, nor to argue that pure assumption has no place in science, but to point out as a matter of practical procedure the utter necessity of adopting the doctrines of the Bacons and the practice of Galileo. Thus with Galileo's work began the

most critical stage of the history of science—the birth of the scientific method.

If time permitted, considerable emphasis would be given to the history of this period, particularly to show that man had at last provided himself with the one effective method for developing natural laws. There is little reason to dwell on it here except to point out that the same human characteristics that were responsible for the opposition to the new thought, are in evidence even today. We seem endowed with a profound conviction of our ability to see clearly and to reason correctly, while we oppose bitterly any attempt at critical verification of our observations or dispassionate check of our conclusions. It is due to this human failing, for one thing, that the scientific method is so misunderstood, even deprecated, except by those who have mastered its philosophy, and for another that the path of science has been so tortuous. By way of compensation it is a happy circumstance that we never hesitate to check the other fellow's conclusions and this questioning attitude has had a healthy effect.

Galileo's particular contribution to the scientific method was the idea of securing sound data and checking assumptions before applying the reasoning by means of which conclusions are reached and generalizations are established. Thus he placed the correct weight on the validity of the premises. The significance of Galileo's innovation becomes clear when it is recalled that except for this one point, alchemy made use of precisely the same processes and by perfectly logical reasoning produced a well rounded out branch of "science". In fact the phlogiston theory "explained" the processes of oxidation and reduction and was accredited a high degree of plausibility when the predictions of theory were actually verified by experiment. The fallacy of this alchemistic doctrine was shown by Black's epoch making experiments which had as their objective, not the correctness of the reasoning, but the correctness of the basic assumptions.

What alchemy provided was simply a mnemonic device and the fallacy of the whole Aristotelian system lay with the failure to verify assumptions. The correction of this situation was the first great contribution of the Renaissance to the scientific method while the second was the utilization of the methods of inductive reasoning. A revolution was thus brought about in the methods of science. It has been well said that the most important invention was the method of making inventions and we may here paraphrase that cogent remark by

saying that the most important discovery of science was the scientific method. Let us now see how this method was used in the chemical sciences, and finally in metallurgy.

Of the purely descriptive branches of chemical science we may quickly dispose. This applies to the studies of the nature of chemical change and to the common reactions such as combustion, oxidation, reduction, and acid attack on metals. This also applies to the nature of air, water, acids, bases, and salts which were early investigated, and to the identity of the chemical elements. The conception of atoms and molecules took a rational form with Dalton's atomic hypothesis which, when developed further as the periodic system of Mendeleeff, was recognized as one of the grand principles of nature. Metallurgy was affected by this work in many and obvious ways but the generalizations which were to raise it to the status of a science came from the development of physical chemistry.

For over two thousand years energy went unrecognized as an important attribute of matter. It had long been known that fire produced heat and the conception of temperature, in some form or other, was universally held. These heat and energy effects were "explained" by the Greeks with their atomic theory and by the alchemists with their phlogiston theory, but the concept of energy was utterly beyond the reach of those who relied solely on observation and reasoning. Thus it remained for controlled quantitative experimentation to give the conception of energy as an attribute of matter. Out of this came the concept of equilibrium which later developments showed to be essential to an understanding of chemical systems in reaction or physical systems in change.

Beginning with the eighteenth century this work was diversified along several lines, according to the interests of the individual workers and with varying success according to their abilities. A great number of chemical reactions were studied and upon being classified it became obvious that many reactions were similar in nature and were following a common pattern. It was only natural that this should attract attention. While many reactions were observed to go to completion, an important group was found which stopped at some intermediate point, or equilibrium condition, at which the reaction was at rest or towards which it was progressing. Work in this field proved to be particularly fruitful and led to the generalization of Guldberg and Waage which is known as the law of mass action. According to this law the rate of reaction is proportional to the

active masses of the reactants. The condition of equilibrium depends on or can be expressed in terms of a characteristic ratio of the concentrations of reactants and products.

The law of mass action was extremely helpful in understanding chemical reactions but it was subject to certain restrictions. In some cases difficulties arose on applying the law because the true molecular species in reaction could not be determined and in other cases, involving solids, on account of the occurrence of phases of variable composition. Even liquid solutions gave trouble on account of differences in behavior between dilute and concentrated solutions, while the case of two liquid solutions required the special treatment of Nernst's "partition law". Chemical reactions of the decomposition type seemed to come in a category of their own and were particularly difficult to understand. Examples are the decomposition of calcium carbonate and of ammonium carbonate which, though they have much in common, are seen to be different when their equilibria are closely examined. The most that can be said is that some of these cases could be treated with the law of mass action by making certain appropriate assumptions.

In the face of a huge accumulation of data it was clear that no one had yet developed a method by which equilibria could be treated from a common point of view. The theorem of Le Chatelier was an attempt in this direction but, while sound, it did not provide the answer. Attempts were also made to apply the new thermodynamics to the study of chemical systems, including Carnot's conception of entropy, though without results of immediate use. It has been explained by one commentator that the conception of entropy was too difficult to grasp and that statement may well characterize the whole situation. It was still necessary to find some fundamental principle which would clarify and order this field and take it beyond the limits of direct observation. This was the contribution of Willard Gibbs who was the first to distinguish between reactions which occur in homogeneous or single phase systems (as in water solutions of two or more chemicals) and those in heterogeneous systems between two or more phases. He showed that these two cases must be treated separately and his major contribution was to an understanding of those laws which govern the equilibria of heterogeneous systems. This was developed as the "Phase Doctrine" which, with the Phase Rule, revolutionized the conceptions of reactions and equilibria.

Where others were bewildered by the great variety of reactions

with no apparent correlation, Gibbs concentrated on the state of affairs at equilibrium and saw that certain relations must obtain in a system at rest, no matter from what direction or by which route the equilibrium was reached. The reaction or change must be only the path by which the final state was reached. Looked at in this light the change may be physical or chemical and Gibbs showed that the conditions which applied to chemical equilibria also applied to such apparently unrelated systems as the solidification of rock magmas, crystallization from solutions, the freezing and boiling of water or of solutions, the evaporation of solids or liquids, the transformation of allotropic substances, and the solidification and transformation of the metallic alloys. However, Gibbs, who occupied the chair of mathematical physics at Yale University, contented himself with the development of the fundamental principles and with the statement of the Phase Rule. Due to a peculiar combination of circumstances his work went unheeded for years by those who could have profited by it. One worker of the period was actually aware of Gibbs' paper in the Proceedings of the Connecticut Academy of Sciences but failed to discover the Phase Rule in it. A glance at that paper makes one highly sympathetic with that scientist.

Gibbs is to be credited with two assumptions which are of the deepest significance to an understanding of heterogeneous equilibria. The first was that the condition of equilibrium between two or more bodies or phases depends on the number, and not the kind, of components of which those bodies are made up. This assumption corrected orthodox thought and with one stroke took the emphasis off the particular elements, molecules, or compounds of which the system was composed. The "number of components" was really the *least* number of independently variable constituents of the system and must be ascertained in a special way. With alloys the number of constituents is simply the number of metals or alloy forming elements. This is simple enough but a better conception of Gibbs' treatment may be had by examining one of the cases which had previously given so much trouble. When studying the decomposition of a compound such as calcium carbonate, it is not simply the number of elements which counts nor even the number of molecular species— CaCO_3 , CaO , and CO_2 —but, of that number, it is only the number which are independently variable. In the case of calcium carbonate the "number of components" would be two, for the other one is related invariably to the first two by the equation $\text{CaCO}_3 = \text{CaO} +$

CO₂. Simple as this principle is, let us remember that it required the most profound thermodynamical reasoning to show its place in chemical and physical equilibria.

Gibbs' next assumption had to do with variations of the equilibrium condition of a system which are caused by changes in the three factors—temperature, pressure, and the concentration of the components in the phases. To illustrate this with a one-component system, ice and water form a stable system at 0 degrees Cent. and atmospheric pressure. This system remains stable under a certain restriction if, for example, we increase the pressure. The restriction is that the temperature must be simultaneously lowered and by an amount which depends in a fixed way on the pressure change. This system was said to have one degree of freedom because it requires fixing only one variable factor (in this case—temperature or pressure) to establish the conditions of both temperature and pressure under which the ice-water system is in stable equilibrium. If we have an equilibrium of three phases—ice, water, and water vapor—neither temperature nor pressure can be changed and still maintain the equilibrium, for such a change causes the disappearance of one or two of the phases. We can say for this case that the addition of a phase to the ice-water system adds another restriction or reduces the degrees of freedom to zero. This means that these three phases can co-exist in stable equilibrium at only one temperature and one pressure, though the values of those two variables at which the three phases co-exist in stable equilibrium would have to be ascertained experimentally. If we have just one phase, like water or ice, temperature and pressure can be altered independently of each other, and hence either phase by itself has two degrees of freedom. Water is, of course, a one component system but multi-component systems were treated in the same way with the provision that the addition of a component adds a degree of freedom. To summarize, a system in equilibrium is either non-, mono-, bi-, or multi-variant, with 0, 1, 2, or more degrees of freedom respectively, depending on the *number* of the variable factors which, when arbitrarily fixed, establish the conditions of temperature and pressure and the concentrations of the components.

This conception of variability recognizes that changes in temperature, pressure, and concentration of the components in the phases, are effective in altering the equilibrium. Specifically, however, similar possible effects of capillary, surface tension, magnetic, and other

forces were eliminated from consideration. This is a justifiable restriction for the three factors included are the important ones and deserve consideration for themselves.

How were such matters to be treated formally? Gibbs' reasoning led to the assumption that the "degrees of freedom" or "variability" of a system in equilibrium were fixed by a relation between the number of components and the number of phases in which they occur. This came from reasoning which was based on the second law of thermodynamics and which started with the assumptions that each phase was individually in a stable condition and that, simultaneously, the free energy of each constituent was the same in all the phases. A simple mathematical treatment of the various simultaneous equations which give effect to these assumptions results in the statement of this fixed relationship which is known as Gibbs' Phase Rule:

$$F = C - P + 2$$

where F is the variability or number of degrees of freedom, C is the number of components, and P is the number of phases.

This Phase Rule has both direct and indirect application to our work on metal and alloy systems. Its immediate use is in a study or examination of typical equilibria to see, for example, if they are limited to fixed temperatures and compositions or ranges and which, by negative inference, would show they are not stable when observed to persist outside such limits. In a broader way it was the availability of the Phase Rule that made it possible for Roozeboom to develop the Phase Doctrine and the laws for establishing and interpreting the constitution diagram. On the other hand the Phase Rule has definite restrictions and, due to failure to grasp the implications of those restrictions, it has been criticized in an unwarranted fashion. In addition one must be cautioned, as the teacher cautions his students, that alloys are seldom in a state of true equilibrium and that the Phase Rule does not provide a code for understanding heat treatment and the like.

Considering the subject matter and the scope of this generalization, it represents inductive reasoning at its best. If it is true that the really great is also simple, the utter simplicity of the principle set forth in the Phase Rule marks that generalization as one of the greatest in science. It is basic to all heterogeneous equilibria which come within the scope of Gibbs' assumptions, and brought order to that field for the first time; situations which were hazy became clear; analogies which escaped attention were brought out; what was a

very complex and hence difficult field became simple and understandable. As metallurgists, we may note that in no other field has the phase doctrine been more useful than in the clarification of the metallic alloys.

I have already noted that Gibbs' work was published in an obscure and inaccessible source. It was also clothed in language that made difficult reading and Gibbs himself apparently made no attempt to apply his principles. The result was that for years it remained unused, even unusable in the ordinary sense. Jumping from the United States to Holland, there was to be found there a brilliant school of Dutch physical chemists at that time, headed by van der Waals and van't Hoff, whose attention was attracted to Gibbs' work. In due course Gibbs' treatise was translated into German by Ostwald and at a later date Professor Bancroft of Cornell University published his own book on the Phase Rule. Thus after about twenty years Gibbs' work became broadly available to physical chemists. The foremost advocate of the phase doctrine, if I read history correctly, was a younger member of that Dutch school, Professor Bakhuis Roozeboom, who for many years studied the equilibria of single and multicomponent systems. An account of this work is brought together in his monumental treatise, "Heterogeneous Equilibria from the Standpoint of the Phase Doctrine", to translate the title.

Roozeboom's career was one of intense specialization in the Phase Rule and problems of heterogeneous equilibria and constitution. His audience was likewise a special one and consisted mainly of teachers and research workers, with a larger number of students who secured the necessary information and interpretations second hand. Late in his life Roozeboom turned his attention to the alloys and showed us how they too may be understood and their behavior followed in a truly scientific fashion. It is not too much to say that in doing this work it was he who established knowledge of the metals and their alloys on a scientific basis and brought order and direction to that highly empirical field. The contribution of Darwin to biology and Pasteur to bacteriology have their counterpart in the work of Roozeboom in physical chemistry and metallurgy; but before attempting to justify that statement I wish to recall the situation in metallurgy at the beginning of this century.

Certain events which are now recognized as being of the greatest significance to the development of scientific metallurgy had already

occurred. First, and indubitably foremost, was the introduction of metallic microscopy by Sorby about 1860. It later became the most important single tool of the metallurgist and still holds that rating without competition. Sorby is universally recognized as the father of metallography and it is fitting, indeed, that his name was given to the most important structure of heat treated steel. By the year 1900 the microscope was used in all metal producing countries for the control of melting, processing and heat treating. In scientific work Sorby's technique was equally productive and a large body of new factual knowledge had been assembled, thanks to Sorby and those who followed.

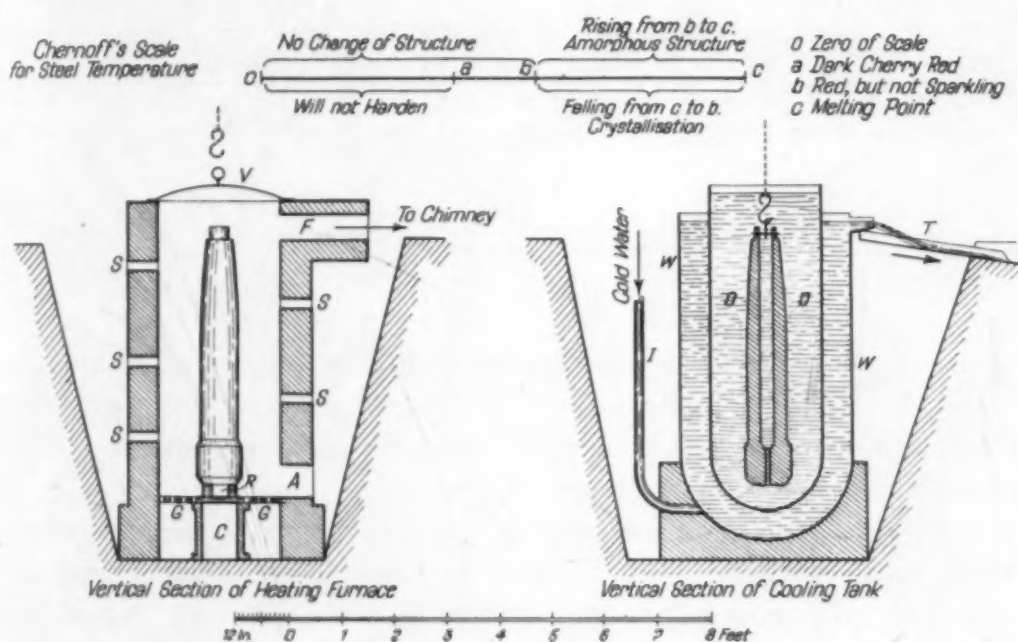


Fig. 1—From Tschernoff's Early Work. Redrawn from Diagram Published in *Mechanical Engineers*, 1891.

At the same time data were being accumulated in allied fields. Tschernoff had observed the "change point" in steel by 1869 and by crude (Fig. 1) though keen methods of observation he had even correlated it with the carbon content. This marked down the first line of the iron-carbon diagram. This study of the transformations of steel was materially advanced by Osmond with his microscopic work and by the use of Le Chatelier's thermocouple. Other experimenters worked similarly on nonferrous metals and alloys and by plotting the various freezing points against the alloy compositions they obtained the so-called "freezing point" diagrams. These depicted the effects of composition on freezing points and also gave additional

thermal arrests, called "eutectics" to which reference will be made later (Fig. 2). Their diagrams were the crude forerunners of our present constitution or phase diagrams.

Engineers and physicists were also accumulating data on alloys and the interesting observation was made that the physical and mechanical properties varied with composition (Fig. 3) in a fashion that seemed to bear some relation to the freezing point diagram. Speculations on this point were supported by microscopical examinations (Fig. 4). Metallurgists, needless to say, welcomed these new

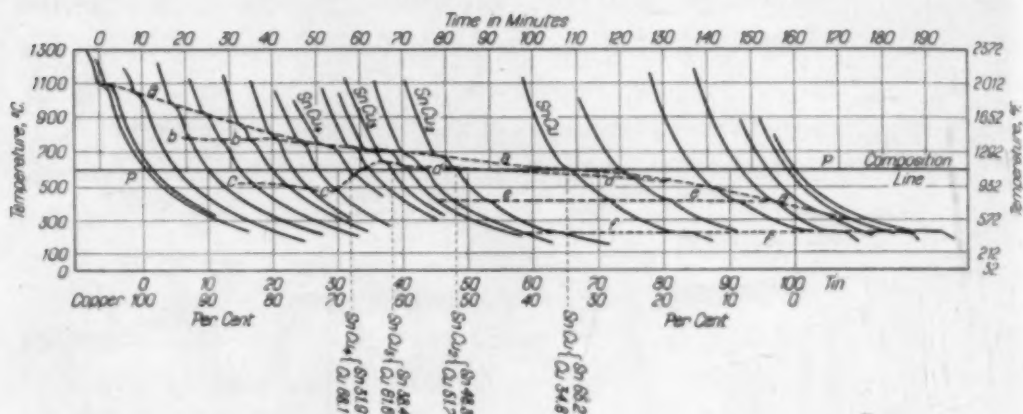


Fig. 2—Cooling Curves of Copper-Tin Alloys. Redrawn from Curves Published in *Mechanical Engineers*, 1895.

ideas and made effective use of them in their various technical pursuits.

Work along these lines was pursued in England, France, Germany, and elsewhere. All this activity brought out a huge mass of data, one of whose significant results was that a very confusing situation was created. To the practical man who was attempting to apply the new findings there were a number of things that were profoundly disturbing in this new metallurgy. One of the leaders of his day pointed out that the photomicrographs were hard to understand and he complained that they made out quenched steel to be akin to concrete, at which he was rightfully indignant and besides everyone knew it was not so. Furthermore they showed only an insignificantly small fraction of the whole piece and to be representative such photographs should be many hundreds of times larger, an accusation that is heard even today. Finally he protested that there were too many "ites" in steel, and the authorities themselves seemed to be in disagreement on them. While these objections might be dismissed as trivial, nevertheless the points were legitimate and

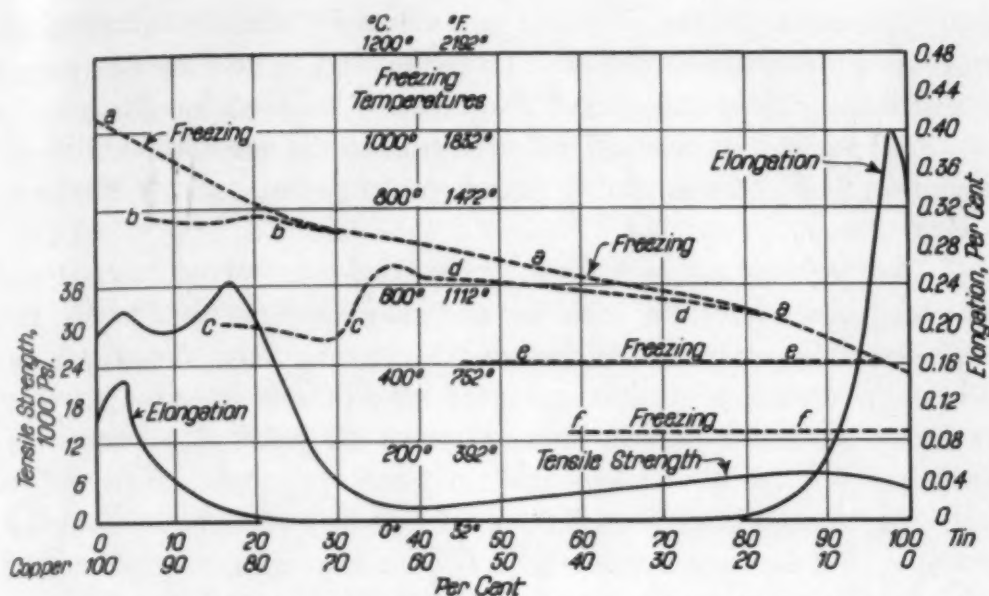


Fig. 3—Tensile Strength and Elongation of Copper-Tin Alloys. Redrawn from Curves Published in *Mechanical Engineers*, 1895.

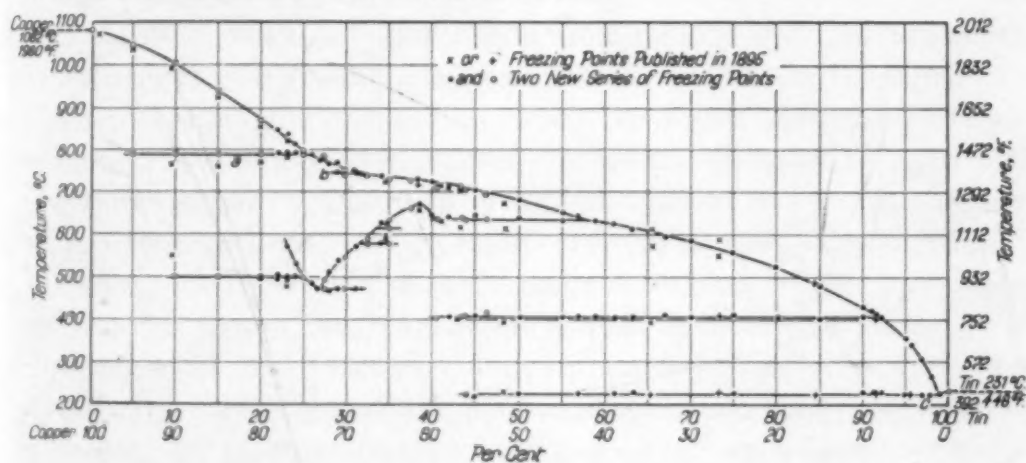


Fig. 4—Freezing-Point Curves of Copper-Tin Alloys. Redrawn from Curves Published in *Mechanical Engineers*, 1897.

brought out some of the minor difficulties encountered in placing metallurgy on a scientific basis. Of greater moment, however, was the inability at the time to understand and interpret the large accumulation of experimental data; thus to the scientist this confusion was far more keenly felt and created for him a typically intolerable situation.

At this stage of its development metallurgy was fortunate in having the stout backing of Roberts-Austen and we may say that poetic justice was again done when this gentleman's name was given to one of the best known constituents of the steels. Possibly his most valuable attribute as a leader of metallurgical thought was a

fine appreciation of the scientific necessities in both the accumulation and the interpretation of data. Unfortunately time does not permit a discussion of his many contributions nor to dwell on his abilities as a leader but I do wish to call attention to his association with the metamorphosis of metallurgy from an empirical to a scientific or ordered basis.

The freezing point diagram occupied the center of interest and was used in studying the behavior and characteristics of alloys. The microstructure and the various correlations already mentioned undoubtedly helped, while the analogous case of the freezing of saline solutions seemed to place interpretations on a more secure basis. On the other hand, alloy systems such as copper-zinc or copper-tin were much more complicated than the simple salt-water system, for example, and this analogy served only as a starting point. The chief difficulty did not arise from the complexity of the problem, however, but was due to complete ignorance of the basic laws and lack of understanding how this type of subject matter is to be handled.

At this time all horizontal lines were called "eutectics", though we now know that they were due many times to other causes. While this may not have been confusing at the time, because the truth was not known, it was at least definitely misleading. Even more misleading must have been the conception of two freezing points. The upper line on the water-salt diagram was interpreted as giving the temperature at which all the excess water froze out as ice and the lower line, or eutectic horizontal, was (correctly) held to be the temperature at which the eutectic froze. No thought was given to an ice-brine equilibrium between those two temperatures or to a continuous formation of ice and consequent enrichment of the brine to the eutectic concentration. The conception of "double freezing points" seemed perfectly logical. Finally the very construction of the freezing point diagrams speaks eloquently enough for the state of the science in those days inasmuch as no attempt was made to close the fields by boundary lines to indicate the regions of one- and two-phase equilibria. While we now know that the early interpretations were incorrect there was then no theory to guide to the correct understanding or even to suggest the necessary experimentation to disclose the truth.

We can readily imagine Sorby asking himself what the compositions were of the two parts of his "pearly constituent" and whether they remained constant in composition and in the same relative

amounts in high and low carbon steels. He might actually have answered those questions correctly, yet without uncovering the basic laws which control such behavior. We can also visualize Tschernoff in the same inquiring mood pondering the change of his point (b) with the carbon content. But it is not difficult now to see how impossible it was in those days to reach a correct or even a helpful understanding of this behavior of steel, or of its constitution.

My object in bringing up these points is to show how confusing it must have been to attempt to understand the accumulated descriptive material no matter with what discernment the observations had been made. Lacking the discipline of a rigid theory even simple inconsistencies went unnoticed, an example of which might be the presence of three phases in a binary alloy. But even without recognizing actual errors a most urgent desire was repeatedly expressed in the literature of the day to get at the basic laws and to apply them to studies of alloy structures and properties. Roberts-Austen was the leader of this movement and in him the urge was especially strong to set matters right. I can do no better than quote Roberts-Austen himself. "It would seem that nature has enshrined in steel a series of her most complex secrets which we must discover as a condition of our successful use of it." It was in precisely such a frame of mind, we may presume, that he requested Roozeboom to contribute his ideas on the iron-carbon system and on a construction of the diagram which would conform to the Phase Rule doctrine.

If we were to attempt to analyze this inquiring attitude, this groping for general and fundamental laws, we might well pose the philosophical question of the ultimate value of such inquiry. Should we not remain contented with the empirical facts, for after all has not the record been that science but explains what is already known and practiced? Do we not hear on all sides that practice is ahead of theory? Let us examine this a moment. Is it not a better statement to say that while science does explain that which came before, it also motivates that which comes after? Scientific inquiry must always deal with things that are known, or with observations and discoveries, but this knowledge must be ordered and understood in terms of fundamental laws before it can be utilized to the best advantage. In the case of metallurgy at about 1900 scientific progress had foundered on the shoal of empiricism and no amount of gathering of data or purely deductive reasoning sufficed to provide the basic laws being sought.

Turning now to Roberts-Austen's appeal to Roozeboom, let us note what happened. With the doctrine of the Phase Rule as his guide Roozeboom had developed the laws of heterogeneous equilibria which are necessary to an understanding of metals and alloys. This was set forth in Professor Roozeboom's paper (Fig. 5) before the Iron and Steel Institute at its Paris meeting in 1900 and entitled "Iron and Steel from the Point of View of the Phase Doctrine". The principles were well illustrated by the work on the iron-carbon alloys and were later applied by other investigators to all other alloy systems which have been studied (Fig. 6). A thorough analysis of

IRON AND STEEL FROM THE POINT OF VIEW OF THE "PHASE-DOCTRINE."

BY PROFESSOR H. W. BAKHUIS-ROOZEBOOM, PH.D. (AMSTERDAM).

It is a great honour for me to present to the Iron and Steel Institute a summary of a recent article of mine under the above title,* at the request of your President, Sir W. C. Roberts-Austen, whose eminent researches form, in fact, the basis for my theoretical study.

My guide in this study is the phase-doctrine of Gibbs. I think as yet this is very little known by most of the members of this Institute, but it is, nevertheless, a conception which has already proved its usefulness in the explanation of many an entangled chemical problem. One of the chief objects of this inquiry is to find out the conditions of the existence of the phases of a system of one or more components.

Fig. 5—From Professor Roozeboom's Paper Before Meeting, Iron & Steel Institute, Paris, 1900.

this contribution to metallurgy would take us deeper into the philosophy of science and the causal relationships of its evolution than time permits. However, we may be forgiven if, forty years later, we take his contribution for granted, for his teachings are now embodied in our formal theories and are submerged in and harmonized with our everyday thinking. He corrected the false interpretations of alloy behavior to which I have alluded and gave a new form and concept to the freezing point diagram, which thereupon became the present constitution diagram; he gave a new meaning to microstructure and it became understandable in terms of the constitution; he

gave new significance to physical and mechanical properties and showed how they are to be correlated with alloy composition and structure. In more specific terms we may say that as a direct result of his teaching we are equipped to handle everyday problems of metallurgy such as melting points and freezing points, dendritic segregation, peritectic reactions, eutectic solidification, multi-component systems, solid solubility and its variation with temperature, transformations in the solid state and heat treatments based thereon, and intermetallic compounds. In brief he gave us not only the iron-carbon and other alloy diagrams but he also founded the methods of interpreting and utilizing them.

Turning to the teaching of metallurgy, as contrasted to its practice, nowhere does Roozeboom's work take on more significance than in the classroom. It has been stated that were we to neglect education for one generation, civilization would stop in its course. I mention this here to point out the function of our schools in the growth of a science and in the utilization of its teachings. Those who are acquainted with pedagogic practice will also appreciate the difference between teaching an empiricism of loosely related facts and teaching a science whose subject matter is properly co-ordinated by basic laws. Thus in the years following 1900 was Roozeboom's conception of the metallic alloys spread to students, and in turn was incorporated in the scientific research of the laboratory and in the technical practice of the industrial plant.

Looking at the growth of scientific metallurgy in retrospect it is clear that the historical development has taken the natural course. Along converging lines came metallic microscopy, pyrometry and thermal analysis, the determination of physical and mechanical properties, and finally the Phase Rule and the laws of heterogeneous equilibrium. When these were united, the synthesis became the modern conception of the constitution of alloys—it was to Roozeboom's labors, let us note, that credit is due for this synthesis.

Upon leaving this phase of the subject it may be profitable to discuss what is meant here by the term "metallurgy". In some circles the word has been used to mean the treatment of ores, though that art is more properly called "ore dressing". In ore dressing the metals are not treated as such, but, rather, as values which must be concentrated or separated from worthless gangue. A second use of the term to mean the recovery of metals from their ores and the refinement thereof is ages old. In more recent years the subject of

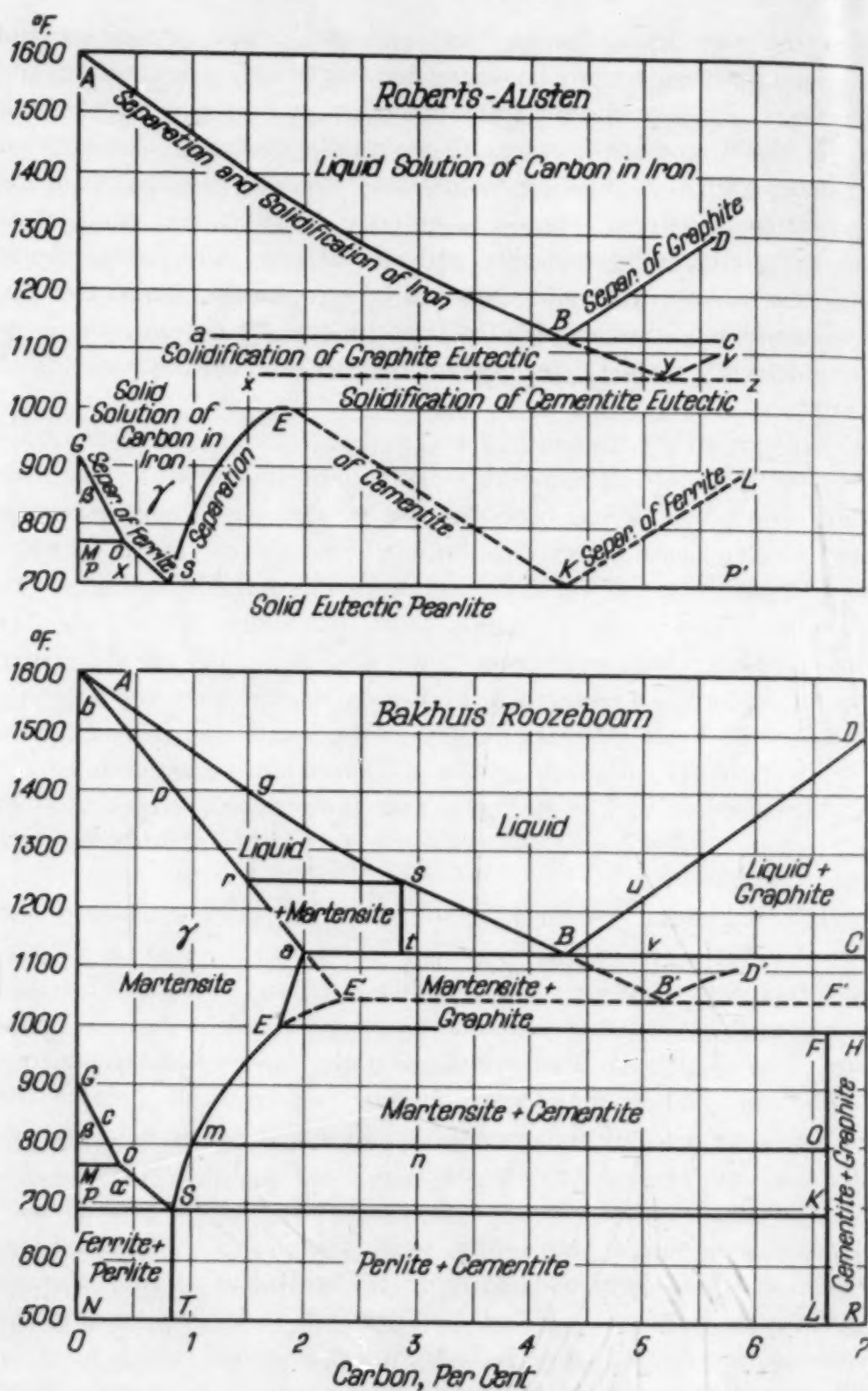


Fig. 6—Equilibrium Diagrams of the Iron-Carbon System by Roberts-Austen and Bakhuis Roozeboom. Redrawn from Curves Published in the *Journal of the Iron and Steel Institute*.

metal behavior has grown so in stature that it has become desirable to distinguish between work on metals—the finished product—and the smelting and refining operations. One way of making this distinction is through use of the terms “process metallurgy” and “product metallurgy”. These terms are more for pedagogical purposes and in general parlance the study of metals is called “physical metallurgy”, or the “science of metals”, or even “metallography” when that term is given a broad meaning. While I do not intend to urge upon you any ideas of terminology I would point out that these various names and definitions are an unhappy source of confusion which has not been resolved. It is perhaps unfortunate that at an early date such a term as “metallology”, or the simpler “metallogy”, was not adopted to connote the scientific study of the metals and alloys. Suffice it to say that this is the field under consideration, and upon leaving the contribution of physical chemistry we turn now to that of physics in another important branch of our subject—crystal structure.

Following Dalton's atomic hypothesis it was generally assumed that atoms and molecules were arranged in crystals according to a fixed and repeating pattern. This is sometimes referred to as the Frankenheim-Bravais lattice. The definite and constant geometry of crystals and the fairly secure atomic hypothesis made this a reasonable assumption. The situation with metals was not quite as clear though it was held that at least some of them were crystalline, the same as minerals. Other metals were held to become crystalline as a result of vibration and repeated pounding, hence presumably were previously in some other condition. Curiously enough steel was once held to become amorphous at a red heat, as a necessary preparation for hardening on quenching. Out of these simple beginnings has developed the science of crystal structure.

The mathematical possibilities of handling the various ways in which atoms could be arranged in space to produce solid bodies were grasped at an early date. A formal statement appeared as the theory of space groups which listed 230 groupings in 32 major classes of the 6 crystal systems. The feature of these groupings was that each one was capable of filling space by parallel movements in three different directions. This theory was mathematically rigid and made it possible to treat crystal structure formally. Out of it came a whole system of crystallography all of which was possible theoretically and inherently consistent, yet pure hypothecation.

For many years the work on crystal structure scarcely affected metallurgical thought. While the crystal symmetry of a few native metals was determined, it was not ordinarily possible to secure metallic crystals for the measurement of interfacial angles. Furthermore the optical methods which were used with such success on transparent materials were not applicable to the opaque metals. In brief, no really effective tool was available.

On the other hand metallographists soon became convinced that metals must be crystalline. This came from observations of dendritic growth of grains, Widmanstaetten figures, etching pits, and the like, and was consistent with the fixed freezing points of metals and the evolution of heat during solidification. The only results of important bearing on metallurgy were the observations of the crystalline nature of plastic deformation by Ewing and Rosenhain in England and Mügge in Germany. This work attracted only small attention at the time (about 1900) for the urgent problem then was that of the constitution of alloys. Furthermore it was not possible to go beyond simple observation, except by speculation.

If I were to follow the historical development it would be necessary at this place to discuss the contributions of Beilby and Rosenhain to the problem of hardening metals by plastic deformation. I refer, of course, to the role of amorphous metal and the various hypotheses built upon that mythical constituent, but it will be profitable to defer such discussion until later. The same is true of the "translation" theory of Tammann and the "lattice disruption" theory of Czochralski. All of these were prominent in polemics on the physics of plastic deformation, until the new tool of X-ray crystal analysis reformed thought in the field.

In the early part of this century there was considerable discussion about the spacing of the atom planes in a crystal lattice, on the one hand, and about the character of X-ray radiation and its wave length, on the other. The turning point came when von Laue suggested that the interplanar spacing of crystals is probably about the correct order of magnitude to diffract the X-ray beam, provided X-ray radiation is the same type as light except for the shorter wave length. Upon having his idea checked experimentally, that assumption was found to be correct and a step was taken which ranks with the great occasions in the history of science. That was in 1912. Many names are associated with the work on crystal structure but the two Braggs, father and son, deserve special mention. They were

the ones who gave it impetus and guidance, beginning in 1914, and they have made important contributions ever since. As I have intimated, this field can be divided into two major subdivisions—the structure of metals and alloy phases, and the effect of deformation on the structure.

The structures of most of the elements are now known, many to a high degree of precision. Upon comparing the structures with the periodic system it is clear that the atomic arrangement recurs in vertical groups of the elements in a typical periodic fashion. Elements whose metallic character is pronounced have structures which are either cubic or the hexagonal close packed. Chemical theory teaches that the forces between the atoms are simple metallic or electronic linkages, with ions at the lattice points and the free electrons more or less free to move in between. Outside of the periodicity just mentioned it is not easy to discover specific chemical effects of the metal atoms in these groupings and the structure seems to be controlled by predominantly geometrical considerations. The metalloid and nonmetallic elements show a lower degree of symmetry and are less closely packed. The interatomic linkage is of the chemical type and these elements follow the rule that the number of close neighbors of an atom is equal to $8-N$, where N is the Group number in the Periodic System or the number of valence electrons. Thus the difference between metallic and nonmetallic elements is shown by differences in atomic arrangement and, more fundamentally, by differences in the freedom of movement of the electrons and the type of linkage. Some of the elements apparently seek to complete their outer shells, similarly to atoms in compounds, by sharing electrons with the correct number of near neighbors.

A structure problem which follows immediately is that of isomorphism or solid solution formation, and upon taking this up it seems desirable to clear up a mistake in nomenclature. When I was introduced to this common constituent of the alloys it was labelled "mixed crystals", which was translated from the original German. Coming directly after a discussion of mechanical mixtures, this name created no particular difficulty, or so it seemed at the time. While this structure had the appearance under the microscope of a pure metal, I reasoned, the individual crystals of the component metals must be on a sub-microscopic scale. By some process, upon which I would be informed later, such an intimate mixture of crystals must behave quite differently from the microscopic mixture of the eutectic.

Of course my concept was false and the anticipated clarification never materialized. It was not until I studied the same subject matter in German that I got the correct picture, for in that language these bodies are called "Mischkristalle". Then I learned that this was a matter of miscibility, not of mixture, and that the atoms of the component metals presumably joined each other indiscriminately in building up the crystal. I presume that experience did me a good turn for it made me suspicious of words, and particularly critical of the misuse of words. This obvious error in translation, by which I mean the unhappy choice of an English term which, in its context, means the exact opposite of the original, seems still to be harbored in our language.

A review of the work on the structure of solid solutions again shows the fruitfulness of the experimental method when properly guided by sound theory and adequately implemented. X-ray evidence showed at once that two types of solid solutions can be distinguished. It has been observed for solid solutions of carbon in iron that the density does not agree with that calculated from X-ray determinations when the assumption is made that the carbon atoms occupy regular positions on the iron lattice. Instead the actual density is somewhat greater and by the correct amount if the carbon atoms are in the interstices. This type of solid solution is formed by small atoms and is called the "interstitial" solid solution.

The substitution type is the more common one with atoms of the added element substituting for atoms of the parent metal on the lattice. In some cases this substitution causes relatively little effect on the lattice while in others the new atoms may be sufficiently different in size or in force fields to produce a noticeable expansion or contraction. When the effect is small the alloys follow Vegard's law and the parameter varies linearly with composition. In some instances the parameter is greater but more frequently is definitely less than that calculated from the rule of mixtures. In the latter case the force of attraction of unlike atoms is manifested.

These and other considerations make it possible to recognize various factors which govern solid solution formation. They are the metallic character of the elements, the structure of the parent metal, the structure of the added metal, the relative size of the two atoms, or the "size factor", the positions of the elements in the Periodic System, and their valences. The precise effect of the added element has been given considerable attention and, along with the

findings to date, future work should enable us to understand solid solutions with high precision. One or two points will be mentioned to illustrate the effectiveness of the scientific method in this field.

The depressions of the freezing point and the solid solubilities of various elements in a common solvent have always appeared to be exceedingly capricious. Hume-Rothery, who has contributed in spectacular fashion to a better understanding of metals and alloys, has shown the great significance of size and valency factors in controlling both effects. With favorable size factors the liquidus curves and solid solubilities in silver of the four adjacent elements, cadmium, indium, tin and antimony, practically coincide when concentrations are plotted as atom per cent \times the valence. In other words the larger the number of valence electrons, the greater the effect of the element in lowering the melting point and the fewer the number of atoms which can find stable positions on the parent lattice. With the consequences of this we shall not be able to deal, but evidence of this type gives a high degree of probability to the picture that is being formed of solid solutions.

In even a cursory review of crystal structure the important case of "super-lattices" must be mentioned. Copper and gold are both in Group I of the Periodic System and form a continuous series of solid solutions. The structure of this system was first studied systematically by our former President, Dr. E. C. Bain. The lattice parameter varies from one metal to the other in a perfectly orthodox fashion but the X-ray films of certain alloys had extra lines in addition to those produced by a simple face-centered cubic lattice, and were called "supernumery" lines. It was reasoned that an ordered arrangement of the gold atoms would give a gold lattice of larger spacing which would produce a pattern of the kind observed. Nothing of this kind is observable under the microscope though Tammann had postulated this type of arrangement to account for the parting limits of the gold-silver and other alloys. Since then other systems have been found with super-lattices and the case is well documented in the literature. This field has proven to be particularly fruitful in giving a deeper insight into alloy structure and a better understanding of the reciprocal behavior of different kinds of atoms on a common lattice.

When the size factor is favorable and the metals are similar and of the same valence, it is common to find complete miscibility. Of course even with all these factors favorable the two metals should be

of the same crystal habit. If they are not, each one dissolves the other to a large extent and a relatively small gap separates the two end or primary solutions. At the other end of the scale come the cases of limited solubility in which the atoms differ too greatly in size or in character to be able to substitute efficiently for one another. Furthermore if the metals differ markedly in character, an extremely interesting body forms, known as an "intermetallic compound", to which we now turn.

For many years it has been recognized that metals form compounds with other metals and with nonmetallic elements. These compounds defied all efforts at explanation on any rational basis and in chemistry they remained a law unto themselves. Calling them compounds at once raised the question as to their possible relationship to the more common compounds of inorganic chemistry, though such comparisons but served to confuse. For example, while we are quite accustomed to H_2O , $NaCl$, etc., in which the elements combine according to their chemical valences, the metallic compounds, such as Fe_3C , $CuAl_2$, etc., were completely baffling because those atom ratios did not follow any known rule or law. The well known Cu_4Sn may have provided some comfort though that would have changed to mild consternation had it been known then, as it is now, that that compound is actually $Cu_{31}Sn_8$.

In the light of present structure theory and the new concepts of interatomic linkages it is possible to bring a highly satisfactory degree of order to the whole field of compounds—inorganic, organic, metallic, and mineralogical—and in this process the simple rule of combination by chemical valence is given the more modest role of a special case instead of being regarded in the time-honored sense as the general rule. In this development the metallic compounds have played an important role.

The past fifteen years of crystal structure have been full of romance but unhappily time does not permit me to do more than mention a few of the more important findings. The significance of this to the scientific method is that implementation was provided by the new X-ray technique. The exact, quantitative data have by this time stimulated thinking on the basic laws governing the formation of alloy phases, and on a classification of the different types of "linkages" or atomic bonds which unite and hold the ions, atoms, or molecules in the lattice positions.

In the common compounds of inorganic chemistry the linkage is

the "heteropolar" or "electrovalent" type, such as is represented by the typical ionic compound, NaCl. The sodium atom gives up its one outer or valence electron to a chlorine atom, whose outer shell of seven electrons is thereby raised to eight, and completed. This giving up and acquisition of an electron produces an ion of sodium with a plus charge and an ion of chlorine with a negative charge. The atom ratio of this type of compound naturally corresponds to the normal valences of the combining elements. In organic chemistry the characteristic linkage is the "homopolar" or "covalent" type and is well illustrated by the diamond structure. The carbon atom has four valence electrons in its outer shell and, of course, requires an equal number to complete a saturated shell of eight electrons. In the diamond each atom is surrounded by four equidistant atoms in tetrahedral arrangement and it appears that bonding is accomplished by each atom "sharing" electrons in pairs with each of its four near neighbors. This union by sharing electrons is common in organic chemistry and we may note with considerable satisfaction that the picture of molecular and crystal structure held by our friends, the organic chemists, is consistent with the new findings. These compounds commonly obey the 8-N rule.

The linkage that is of greatest interest to the metallurgist is the "metallic". The typical metal is pictured as built up of metal atoms or ions at the lattice points with valence electrons in the "unoccupied" space between the ions. These electrons behave somewhat as an "electron gas" and under some conditions may even pass from one atom to another. It is they that provide the cohesive forces and no doubt this electron mobility is related in an important way to the ability of metals to deform by slip without disrupting the interplanar bonds. The free electrons are also the source of metallic conduction, because they are free to move, and they give the typical reflectivity.

The intermetallic compounds have been very prominent in developing these concepts and, in turn, cannot be understood on any other basis. This field of metallurgy has now been beautifully ordered and while it would not serve our present purposes to discuss it in detail, one case will be cited to show the effectiveness of the scientific method in dealing with this difficult situation.

In alloys of the heavy, high melting metals with certain low melting metals we find phases of similar structure recurring. There is something about tin when added to copper and about cadmium when added to silver, to cite two instances, that produces the same

sort of intermediate phases. Even before the days of structure analysis Professor Carpenter suggested that there must be some correlation between corresponding phases of similarly shaped diagrams and from X-ray structure data his suggestion is now known to be correct. Though at first glance it is not clear why this should be so, Hume-Rothery has given the key to the situation by showing that the structurally similar phases of systems such as copper-tin, copper-zinc, silver-cadmium, etc., have the same ratio of total number of valence electrons to total number of atoms. This is now recognized as one of the fundamental relationships which control the formation of an important group of intermetallic compounds. The ratios are 3:2, 21:13, and 7:4, and the experimental findings agree with this postulate in so many cases that this ratio, rather than the relation of the valence electrons, seems to be the controlling factor. One such phase is the $\text{Cu}_{31}\text{Sn}_8$ mentioned earlier which X-ray data and structural considerations show to be of this composition and not Cu_4Sn . It has 63 valence electrons to 39 atoms, or 21:13. These "Hume-Rothery" or "electron" compounds are of great scientific interest for they offer a clue to one of the important types of chemical combination.

Other compounds are recognized as being dependent on such factors as atom size, number of outer electrons, and Group and Period in which the components occur. The surprising failure of the metallic compounds to follow ordinary rules of valence may be said to be no longer a problem. In this connection I wish to quote from Hume-Rothery as follows:

"In the ordinary compounds of inorganic and organic chemistry the metallic properties are lost in the compound because the loosely bound valency electrons of the metallic atom are bound into the stable groups of the ions or covalent linkages. The usual valency rules are then the result of the fact that certain definite numbers of electrons constitute stable groupings, and that the whole of the available valency electrons are bound into the stable group. Consequently, if free electrons are left over to form a truly metallic compound, the valency relations must in general be different from those in the normal compounds where all the valency electrons are used up, and if the compound is really metallic we shall in general expect the ordinary valency rules to break down."

The great variety of compounds, both metallic and nonmetallic, comes from the variation from element to element of the factors

which control miscibility and compound formation. Apparently exceptions are due to our lack of understanding how to weigh the factors or to some factor not yet recognized.

It is hoped that this hasty sketch of crystal structure and of the newer concepts of the metallic compounds will suffice to indicate the important advances which have recently been made in that branch of our science. Speculation on these matters was not able to accom-

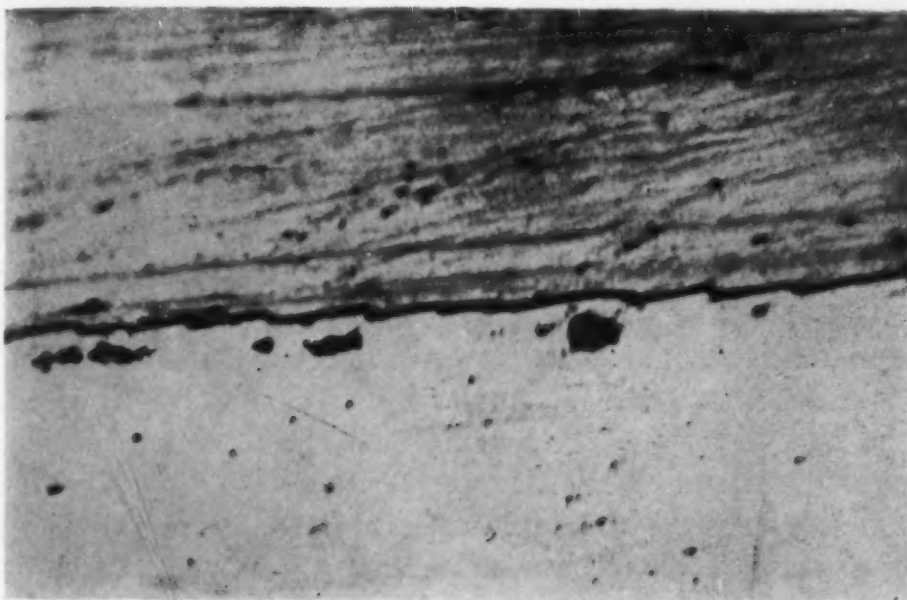


Fig. 7—Section Through Electroplated Piece of Iron, Which Had Been Previously Polished and Strained, Showing Minute Slip-Band Steps About 0.00002 Inch. Magnification 1200 Diameters. (Rosenhain)

plish as much in twenty centuries as have the methods of X-ray crystal analysis in twenty years. The Greeks, we may note, could scarcely be expected to make any useful contribution with their "atomic theory", even with their unparalleled facility in the processes of reasoning. In recent times even with the atomic theory of Dalton, the theory of space groups, and the valency theory of chemical combination, no general conception of structure or of chemical combination was forthcoming. This triumph of the scientific method is, in a sense, a fitting introduction to the problem of plastic deformation for it is in this same way that our ideas of that process should be built.

In establishing a theory of plastic deformation the first point to be settled is the precise mechanism of the movement or displacement of metal, yet it appears that too little attention has been paid to this

point in recent years (Fig. 7). Ewing and Rosenhain found, among several possibilities, that metals behave as though made up of a large number of minute blocks and that during deformation these blocks slip over one another in response to the applied stress. This sliding action did not sever the bonds between the blocks and after deformation the metal was apparently as sound as before. Directly following this original work Rosenhain applied a skillful technique to a further study of this problem. By protecting the surface of a sample which showed slip bands by coating it electrolytically with copper, and then sectioning it, he demonstrated the serrated "edge effect" that would be expected from the simple block movement just described. He also secured evidence of nondistortion of the blocks by showing that the same blocks, photographed before and after an increment of slip, showed no stereoscopic effect. In other words his picture was simply that of a deck of cards with the edge of one card extending slightly beyond its neighbor. Supporting evidence of a simple block movement has been given a few times in the literature, and Andrade, for example, mentions the laying bare of a strip of the slip plane 0.1 mm. wide as one block slipped past its neighbor. This simple picture has survived throughout the intervening years and, with the assistance of models of deformed single crystals, it has become implanted in the minds of many as *the* mechanism of slip. Of course the picture we have in mind of this process must be tremendously important to a theory and while I do not suggest that Rosenhain's picture is incorrect, it is assuredly not the only one and I shall point out later that a fundamentally different and more general mode of deformation has received too little attention, commensurate with its importance.

In the meantime the searching methods of X-ray crystal analysis have been applied to the problem of plastic deformation and test pieces in single crystal form have been used to gain a simpler and more perfect portrayal of the fundamental phenomena. As a result considerable information is now available on the stress-strain relationships and on planes and directions of slip. Actually the data have piled up like logs in a jam and the present situation corresponds to that in metallurgy at the time Roozeboom clarified the question of the constitution of alloys. The king log, in terms of an adequate theory of the metallic state, has not been found and we do not yet know how to classify and order the data or how to interpret them in terms of general laws. By this I mean particularly that the precise atomic mechanisms of plastic deformation and strain harden-

ing are not understood. Nevertheless many competent scientists have tackled these problems and it is significant that the metallurgist has given way to the physicist and metallurgical physicist. Our principal fear as metallurgists is, I believe, that we may not be capable of following the reasoning or of grasping the significance of the results, though we may trust that some intermediary will arise to give us a simplified version of the theory. Consequently it is with considerable diffidence that I venture to point out certain features that seem to require special attention.

On reading current contributions to the problem of plastic deformation it appears that a great need exists for close co-operation between metallurgists and physicists to develop a clear conception of just what it is that must be explained. Instead we find that one starts with one picture and another with a different one, yet each is covering the field and attempting to develop a general theory. The terminology also is highly confusing and various names are used to signify what is presumably one and the same thing. Thus Ewing and Rosenhain showed with the microscope that "blocks" form, and X-ray crystallographers have equally good evidence in their diffraction patterns that the grains develop "fragments". It is particularly confusing to have both discussed under those or other names without a clear indication of which is meant, or even whether the two are one and the same. I would propose, as the first and highly desirable step, setting up a detailed and critical classification of the important phenomena. The object would be to bring everyone in agreement on what is and what is not known, and to suggest what is lacking and what points need special attention. Let us take an example which comes from the work mentioned on single crystals, and which is selected because it bears on our picture of the mechanism of the movement during plastic deformation and on our understanding of the related phenomena.

Single crystals of metals which have one slip plane, like the basal plane of the hexagonal close packed zinc, stretch out in tension as the slip planes tilt more and more into the direction of pull, until rupture occurs (Fig. 8). At the same time the slip planes rotate in their own planes as the direction of slip in the basal planes lines up with the direction of the major shear stress. This gives the simple elemental picture of the geometric behavior of the slip plane and slip direction under the action of an applied stress. It becomes a little more complicated in metals whose structure permits slip on more

than one plane. Thus the face-centered cubic metals have four sets of slip planes—the octahedral planes. If two of these four sets are symmetrical with the axis at the start, the crystal elongates by simultaneous slip on both sets, the details of which need not concern us here. If, however, the slip planes are oriented in some other way or at random, slip starts on only the one set which is most favorably oriented. The extension of the crystal tilts those planes and rotates them as was just described for zinc. Sooner or later, depending on

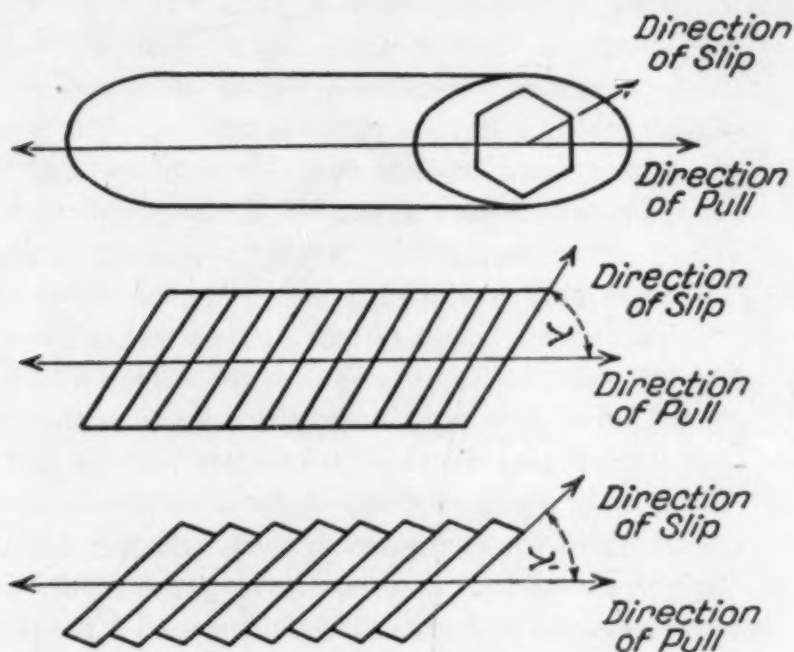


Fig. 8—Representation of Direction of Slip in the Basal Plane of a Zinc Single Crystal. (Elam)

the initial orientation, another set of octahedral planes must be brought into an orientation which is symmetrical with that of the "active" slip planes. Let us see how that fits the picture of a simple block movement. From the standpoint of mechanics the situation is the same as that for the initial orientation which has two sets of slip planes in symmetrical position, by which I mean that the resolved shear stress is the same on both sets of planes. While it will be anticipation I should like to pose the question of what should happen at this point and require that the answer be given on the basis of the picture of simple block movement. Slip on the active planes has presumably strengthened them and we must argue that since no such movement has been going on on the second set, the latter must still be in a virgin condition and hence weaker than the first set. On

this basis we would say that double slip would not occur at the symmetrical orientation, but at one for which the ratio of shear stress to resistance to slip is the same for both active and latent slip planes with the proviso that the latent slip planes still preserve their initial low strength (Fig. 9). This deduction follows from Rosenhain's simple picture of the deck of cards movement. However, the situation is entirely different—double slip starts when the second set of latent slip planes is approximately in the symmetrical position. Pre-

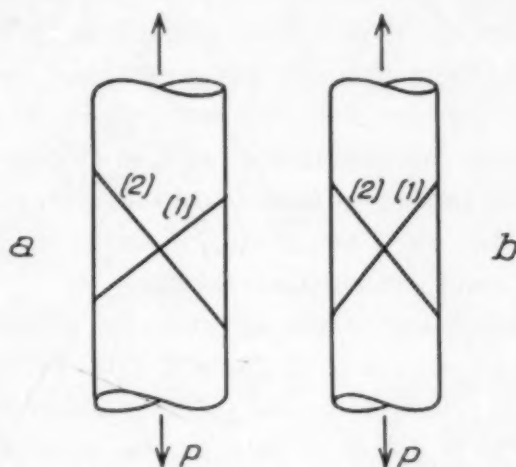


Fig. 9—At Left is Indicated the Initial Positions of Active (1) and Latent (2) Slip Planes in a Face-Centered Cubic Crystal. At Right is the Deformed Crystal at the Start of "Double" Slip.

cise measurements show, in fact, that the latent slip planes which have supported no slip, in the usual sense, have been strengthened by some other though equally potent mechanism even more than were the active slip planes. This is a basic fact of plastic deformation and must be tremendously significant to those who attempt to develop a picture of plastic deformation. It is definitely inconsistent with a process which assumes no change whatever in the blocks proper and which therefore cannot account for strengthening in directions or along planes which cut those blocks at an angle.

In addition to such difficulties of a purely theoretical nature, direct observation shows that metals deform and strengthen by a process which departs fundamentally from the deck of cards picture. This brings us to the pseudo-block movement which I described in 1926 for zinc crystals. As a part of the study there reported we attempted to check the picture of the simple block or deck of cards movement (Fig. 10). A number of critical tests for the serrated

"edge effect" were entirely negative. Slip bands were observed, it is true, but as close as could be observed the visible blocks distorted approximately the same as the crystal did as a whole. No trace of a serrated edge was found though it would have been clearly visible. While this general situation has been discussed in great detail in the intervening years, I know of no reason to doubt the correctness of those observations. This gives two general pictures of the slip process, both of which appear to be well supported. My principal argument here is that the simple block movement is by no means the only one, and I suspect it is relatively the one of lesser importance. Possibly there may be a gradual transition from one type of movement to the other. At high temperatures the slip bands are far apart and the simple block movement is clearly in evidence; at lower temperatures more and more slip bands form and they increase in number as slip proceeds. As a suggestion it seems likely that block distortion sets in at some intermediate stage.

The pseudo-block movement obviates the difficulties in attempting to understand observations of single crystal behavior on the basis of a deck of cards movement, and places much less significance on the visible slip bands either as the major seat of slip or as the source of the strengthening effect. This is somewhat equivalent to saying that our usual methods of observing blocks and slip bands are not adequate to the task, an observation that I pass on to the aforementioned metallurgical physicists. Furthermore, while the slip process appears to be one involving the planes of a crystal, we should not lose sight of the fact that the further the situation is studied the clearer it becomes that the slip direction is of primary, and the slip plane only of secondary, importance.

It is not my purpose here to propose how the physical theory should be developed but rather to point out that these basic facts and purely objective observations should be used as a guide. Furthermore, any generalizations, though arrived at by other means, will have to be consistent with them in all particulars, before they can be accepted. Turning to a brief review of work in this field, the approach of the scientific method has been largely neglected. As an example we may take the work of Taylor who has recently stated and criticized the principal ideas on the mechanics of plastic deformation from the physical viewpoint.

One idea assumes that deformation breaks up initially sound grains into fragments which become disoriented and thus make the

progress of slip more difficult. Taylor held this not to be valid for metals which have several sets of slip planes, such as the face-centered cubic. By appropriate combinations of slip (which we must assume would come about automatically) deformation could take place without producing any geometrical inconsistencies at the bound-

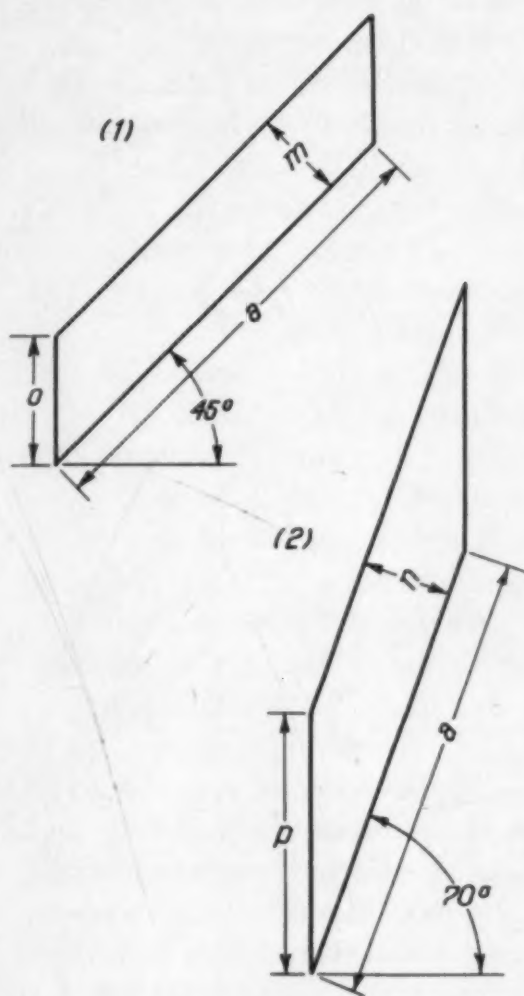


Fig. 10—Basal "Block" of a Zinc Crystal (1) in Initial Position; (2) in Final Position.

aries between fragments and therefore without increasing the resistance to slip. This being so, the strengthening effect would be left unexplained.

A second idea assumes that the metal crystal is initially a mosaic composite of slightly different orientations. It can be assumed that the interfaces carry the entire stress but this would not explain the increase of stress with strain. An alternate assumption might be

that a block can support any stress system which does not have an effective stress component parallel to the slip planes, but this again would not hold for those metals which have more than one set of slip planes.

According to the third idea, to quote Taylor, "A perfect crystal is supposed to be capable of withstanding a very large stress. The observed weakness of metal crystals is attributed to concentrations of stress due to internal surfaces of misfit or cracks, and the increasing strength with increasing plastic strain is attributed to an increase in the number of faults or cracks. As the number of such faults increases, the ratio of maximum stress in the region of stress concentration to mean stress in the material would be expected to decrease. The mean stress necessary to cause a given maximum stress in a region of stress concentration must, therefore, increase as the number of faults per unit volume of the material increases."

Taylor accepts this third idea as the basis of his own theory of "dislocations", which name he uses for the faults or cracks just mentioned. Upon the application of a load or shear stress above the yield point a dislocation is created on one or more of the active slip planes which gives a local alteration of the positions of the atoms. The body of a block or fragment between two such slip planes supports the same shear stress as that at the interface, and it is a point of this postulate that the crystal is assumed to have a high natural strength (Fig. 11). These dislocated atoms along the slip planes are the mechanical embodiment of stress concentrations. By making certain plausible assumptions covering the rate at which the flaws increase in number with deformation, the effects of boundaries, and the effectiveness of the flaws, the calculated stress-strain diagram agrees well with the experimental data. This appears to be verified prediction but Masing, who has subjected this theory to criticism, states it as his opinion that this coincidence is illusory. Masing is well known for his work on internal stresses and their effects on the strength of metals, and he points out that such stress concentrations at faults are capable only of accounting for the Bauschinger effect, but not for an actual increase in the resistance to shear or true strengthening. What the cause might be of the formation of a dislocation is not stated, but at this stage of the theory is of no particular importance. It would be of greater immediate importance to show (which is not done) how the dislocation theory is to be reconciled with the basic facts of plastic deformation. Presumably Taylor is

discussing dislocations on the visible slip planes, but we are bound to ask, for example, by what mechanism the visible blocks deform; how block distortion is to be understood in view of the assumption of great natural strength of the blocks; and how it is that latent slip planes become strengthened.

It may very well be impossible to picture the atomic mechanism of slip in terms of a mechanical analogy, but microscopic and X-ray work shows that the picture of block or fragment formation, in some form, does appear to be legitimate. While there is no proof of it as

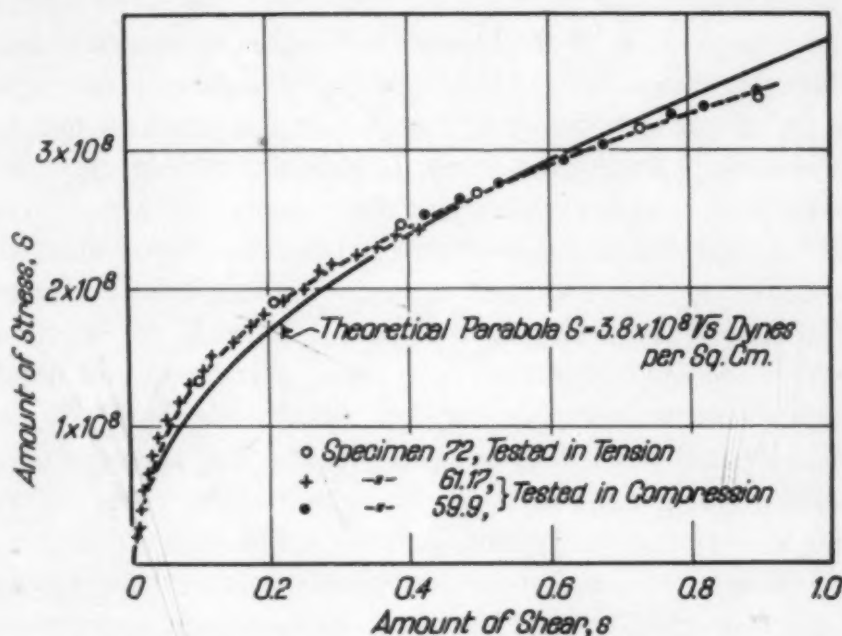


Fig. 11—Relationship (Theoretical and Experimental) Between Shear Stress S and Amount of Shear s for Three Aluminum Single-Crystal Specimens When Slipping in One Plane, According to Taylor.

yet, these structure elements also behave as though relatively strong which might mean that they ultimately form of such a small size that they support the applied stresses by simple elastic distortions. This is consistent with X-ray evidence but before a theory can be held to be adequate it must also answer the old question of why the atoms do not simply and spontaneously revert to their normal positions without the need of resorting to elevated temperature treatments. Numerous facts of atom mobility suggest that the atoms should easily shift the slight amount required to produce straight rows or planes.

I have touched on the hardening or strengthening effect of plastic deformation in connection with the mechanism of the process

but it has been so prominent in metallurgical discussions that it deserves special consideration here. In passing, I would point out that the metal is not necessarily made less malleable, and the ultimate theory of strengthening must be consistent with the retention of a deformation mechanism.

Shortly after the turn of the century Beilby postulated the formation of hard, amorphous metal along the slip planes of Ewing and Rosenhain and it was asserted that this amorphous metal made slip more difficult and the metal stronger. This hypothesis was built up with such skill and argued with such persuasive logic that it dominated thought in this field for years. However, no tests were devised, at that time, by which the postulates could be either proved or disproved and the theory remained pure speculation. Amorphous metal was never shown to exist in the interior of metals, and its properties and characteristics were never under observation. On this account the amorphous metal hypothesis was always open to the eminently fair objection that it was absolutely unfounded, though in saying this it should be kept in mind that the question was not whether amorphous metal *could* form, but whether it *did* form. Just as Black's tests proved the early phlogiston theory to be untenable, so have recent experiments shown the amorphous metal hypothesis to be invalid.

The first critical inspection was probably that of Jeffries and Archer who pointed out that the hypothesis was entirely inconsistent with the behavior of copper and tungsten when those metals were cold drawn by large amounts. However, soon thereafter a crucial test was provided by single crystal work that was particularly potent because it applied a fundamental principle of logic. For example, if it be hypothesized that it is raining outdoors, a simple inspection of the pavements may show that the hypothesis is incorrect. Wet pavements are a necessary consequence of rainfall. Of course evidence can be questioned but the argument soon changes from one of rainfall to one of the condition of the pavement. A situation of this kind gives rise to a valuable crucial test.

In the present case we have to consider not the postulate of amorphous metal forming along slip planes but the validity of the single crystal evidence. In 1926 I drew attention to a certain behavior of zinc crystals which showed that amorphous metal does not form along the slip planes. As a crystal is being stretched in tension it is observed that the blocks of the central portion rotate

spontaneously. It becomes obvious on inspection of this rotation that it is due to a component of the principal shear stress. The slip planes are required by their structure to slip in the direction of a diagonal axis which, in general, is not in the direction of the principal shear stress. One component of this shear stress produces the slip while the other component, at right angles, acts as the force which lines up the slip directions with the direction of the principal shear stress. According to Beilby's hypothesis the blocks would be separated by amorphous metal which had formed during slip and, by definition, would have no directional properties. Actually this rotation is uniquely crystallographic in nature and could not possibly be transmitted from the fixed ends of the crystal in the grips through intervening material having an isotropic or even quasi-isotropic structure. Objections to Beilby's hypothesis have also been raised by Carpenter and Robinson who comment that "As this hypothesis does not offer any explanation as to why the different planes which successively become involved in slip should offer a progressively increasing resistance to it, it does not explain this most important fact". The most direct evidence seems to me to be the point I advanced before this Society in 1936 in the symposium on plastic deformation. With Beilby's picture of strong blocks, it is the well verified finding that the latent slip planes of face-centered cubic metals, along which no slip has taken place, are strengthened even more than were the active slip planes, along which slip did take place. Obviously amorphous metal has nothing to do with this.

My object here is not to bury Beilby's hypothesis, but to point out that it, and other similarly founded metallurgical hypotheses, offer examples of a procedure that should be avoided in scientific work. Amorphous metal occupied a position precisely similar to that of the other mythical constituent of metallurgy, beta iron, which, though its occurrence in high-carbon steel was never proved (and hardness was never shown to be an outstanding characteristic), was held to be responsible for hardening on quenching. In both cases we have examples of the method of Aristotle, and amorphous metal, like beta iron, was simply endowed with those properties for whose explanation it was created, and then assumed to exist at the proper places. That is not the path of science and one is tempted to assert that this theory was doomed to end in failure. That would be unwarranted, of course, but it would be far better if such ideas were held back until proper support of plausibility of the conclusions, but

particularly of the assumptions, can be contributed. In saying this I distinguish between an idea derived from experiment, which produces the body of a theory, and assumption and reasoning, which are used at times as a mere substitute. The latter type of support, as exemplified by assumptions of amorphous metal and beta iron, are altogether too reminiscent of the device used by the Greeks who endowed matter-stuff with whatever attributes were needed to explain their observations. Today this may be light reading but these truly erroneous concepts are befuddling to our thinking and we should challenge the proponents to hold their fire until ready with the proofs. Furthermore a proposition, once it is publicized, is hard to eradicate, and I am tempted to quote the words of the famous poet and philosopher, Goethe.

His friend Eckermann comments on the spreading of false teachings, "It is regrettable that there are so many false teachings, a young man does not know which one to follow". To which Goethe replied, "We have examples of that. We have had whole generations go wrong on false maxims, and suffer therefrom. And then today, the ease with which each error can be broadcast through publication. Even if later the critic sees better and even publishes his better convictions, his false teachings will have taken effect and, like weeds, will survive along with the good".

CLOSURE

I have discussed the contribution of physical chemistry to an understanding of the phase relationships of metals and alloys and of physics to an understanding of the atomic arrangement or structure. We might appropriately have considered work on diffusion, surface phenomena, reaction velocities, nucleation and grain growth, allotropy, electrical and magnetic properties, mechanical and chemical properties and other contributions of the physical and chemical sciences. It is significant that our knowledge of these matters has been largely contributed by investigators who have mastered the philosophy and used the technique of the scientific method. This is a powerful and completely convincing argument in favor of this formal procedure, yet it cannot be denied that other methods might (by chance) be effective too. We are dealing ultimately with facts and truths which conceivably could be arrived at by guessing, assumption, reasoning, by accident, or by the empirical method of trying everything.

Hypothecation, as a formal method of setting up a postulate, is an eminently proper tool of science, though it should be distinguished from inadequately founded assumption. Its successful use, in my belief, can only be hoped for if the investigator has his eyes open and is fully aware of the strength and scope of the speculation he is using. Under such auspicious circumstances the advent of new and better information can be turned directly to account and any incorrect ideas gracefully discarded.

It is well within my own memory that crystals of rock salt were considered to be built up of molecules of NaCl. When the electrifying news came that, instead, the units of structure were ions of Na and Cl, it was a simple matter to correct the earlier ideas. I also well remember the caution with which my professor of physical chemistry put forward the postulates of Arrhenius on the dissociation of compounds in liquid solution so that the line was well drawn between experimental fact and postulate. These are good examples of correct procedure though man has not yet devised a sure-fire method of making postulates. This weakness seems to be particularly in evidence in metallurgy, where theories have been advanced and tenaciously held, though found later not to have been properly founded in the first place. It is truly not simple to theorize effectively, and yet not expose one's self later to adverse criticism. Gibbs seems to have found the best method for announcing a new principle. As stated by Ostwald, "In a strictly mathematical manner, and with a text so concentrated that every page requires the active co-operation of the reader, the author takes us through his 700 equations, only seldom illuminating his results with any suggestive applications". In time this great truth of the Phase Rule came to light and Gibbs has been hailed as a master of inductive reasoning and the greatest American scientist, yet if he had been wrong nobody would have known about it.

While the scientific method gives us a sound procedure for developing the natural and physical sciences, there is as yet no substitute for inductive reasoning by which the man of genius reasons backwards from effect to cause, and on to basic principles or to the correct and appropriate postulate. This was so for the fields of heterogeneous equilibria and crystal structure and other branches of metallurgy which are nicely ordered, but is not yet true for plastic deformation in which field there is speculation, confusion, and disagreement. If I were to theorize on plastic deformation I believe

I would first secure an adequate picture of the macro-, micro-, and atomic mechanisms of the process. I would then make use of other recognized features of the metal lattice to secure a more complete understanding of its virgin condition and the ability of the crystal to distort permanently without rupture. We need not hesitate to accept the role of electrons providing the bonds between the metal ions, while it is reliably reported that the Cauchy relations, at least by negative inference, suggest that we must consider ions and electrons as the structural units. Furthermore, there is good evidence that the valence electrons are more or less free to move and I would want to consider this characteristic when attempting to account for the typical plastic behavior and continuous bonding of metals as contrasted with the viscous behavior of fluids. Finally, and I believe this might well be a guiding principle to all of us, if it appeared that the theory of the metallic state had not progressed far enough for such applications, I would consider that further speculation should stop for the time being until a deeper insight in the problem was obtained.

The dominant note of the teachings of the scientific method is—adhere to demonstrable facts and build postulates in accordance therewith. If, instead, we elect to clothe ignorance in empty words we run counter to the experience in scientific procedure which has been gained through the travail of the centuries since Galileo stood on the tower at Pisa. I am reminded of the caution expressed by Goethe when discussing this human failing.

"Many people are contented", he is quoted by his friend Eckermann, "if they simply have words to use, which my Mephistopheles appreciated and stated rather well: (attempting to deceive the student who seeks his advice on schooling)

'For that which is or is not in the head
A sounding phrase will serve you in good stead;
Above all, to words hold fast,
Then pass through the portal true
Into the temple of certainty;
For there precisely where ideas fail
A word comes opportunely into play'."

THE DEVELOPMENT OF ALLOYS FOR USE AT TEMPERATURES ABOVE 1000 DEGREES FAHR.

BY E. R. PARKER

Abstract

Methods for increasing the resistance of a metal to deformation are discussed. It is suggested that factors which affect the deformation of a metal at low temperatures are also effective at high temperatures, but the influence of temperature on the various factors must be considered in an analysis of high temperature strength.

A critically dispersed precipitate offers the maximum resistance to deformation at both low and high temperatures. At low temperatures, the precipitate remains in the critical dispersion, and high strength is maintained for long periods of time. High temperatures cause the precipitated phase to agglomerate with a consequent weakening of the material. Carbides agglomerate relatively rapidly at temperatures above 1000 degrees Fahr., but certain inter-metallic compounds (e.g., Fe_3Cb_2 and Fe_3Mo_2) are more stable; so higher temperatures and longer times are necessary for agglomeration. An alloy containing a critically dispersed stable compound should possess the greatest high temperature strength.

Rupture tests were made at 1100 degrees Fahr. on precipitation hardening iron-base alloys containing Mo and Cb. The test results indicate that such alloys have greater high temperature strength than alloys which depend upon carbides for strength.

THE utility of metals and alloys for most applications depends primarily upon their response to applied force. Alloys are required to withstand various kinds of forces under many different conditions without undergoing any appreciable amount of deformation or without breaking. The forces applied in service are usually complex, but certain preliminary indications of suitability may often be obtained by testing. At room temperature, the relation between stress and deformation of steel is often sufficient to determine its suitability for use. At elevated temperatures, however, tests related to the suitability of materials are much more complex. Account

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must be taken of the interrelation between stress, deformation, temperature, and time. Further, metals and alloys are not stable at elevated temperatures, but undergo continuous changes in structure with time which in turn affect their resistance to deformation. Structural changes are generally accelerated by deformation; so the relation between structural changes and the factors listed above must also be considered.

It is well known that permanent deformation of a polycrystalline metal involves movement of constituent parts of the crystals with respect to each other along the most widely spaced crystallographic planes and in the direction of the lines in which the atoms are most closely packed (disregarding for the present the effect of twinning). In any cubic crystal there are several combinations of planes and directions of slip, and the imposed conditions determine which will be involved. In a polycrystalline aggregate, each crystal must deform by slip on crystallographic planes and in crystallographic directions, but the metal as a whole must deform as determined by the applied force. It is necessary for each crystal to deform in conformity with the deformation of its neighbors and the metal as a whole, if cohesion between crystals is maintained. Any given crystal in an aggregate may deform as demanded by the change in shape of the metal as a whole without departing from its characteristic behavior of slipping on certain crystallographic planes in certain crystallographic directions by slipping on a sufficient number of slip planes and in a sufficient number of directions. Because the orientation differs in adjacent grains, deformation is retarded by crystal boundaries. Thus a decrease in crystal size with a consequent increase in grain surface per unit volume, will be accompanied by an increase in resistance to deformation.

Another method for increasing the resistance to deformation is to substitute atoms of different size and properties for some of the atoms in the parent metal. A solid solution of this type will require a higher stress to produce slipping because the symmetry of the atomic arrangement has been distorted by the addition of the stranger atoms to the lattice. The size and nature of the solute atoms determine the increase in resistance to deformation. Bain (1)¹ has shown the probable hardening effects of various elements dissolved in pure iron. He has shown that to raise the Brinell hardness of iron

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

about 100 points would require in solution either 30 per cent Chromium, 25 per cent Tungsten, 16 per cent Vanadium, 14 per cent Molybdenum, 10 per cent Nickel, 4.5 per cent Manganese, 3 per cent Silicon, or 1 per cent Phosphorus.

The addition of an element in solid solution will increase the strength of a metal considerably, but a far greater increase is produced by a finely dispersed precipitate. Less than 0.1 per cent carbon when present in iron as a finely dispersed precipitate of iron carbide will cause an increase in hardness of more than 100 points Brinell. Similar hardening is common in nonferrous precipitation-hardening alloys.

It is apparent that there are at least three ways for increasing the low temperature resistance of metal to deformation; namely, reducing the grain size (cold working produces a similar effect), the formation of a solid solution, and the presence of a finely dispersed precipitate. Of the three, the last is by far the most effective.

The resistance of a metal or alloy to deformation at elevated temperatures depends upon the same factors as the low temperature resistance to deformation, but their effect may be greatly modified by temperature and time. The atoms are so immobile at low temperatures that the effect of thermal agitation on the strength may be disregarded. At high temperatures, however, the thermal movement of the atoms becomes so great that they are able to move about in the crystal lattice (as is shown by the diffusion of one metal through another). This movement of atoms is dependent upon both temperature and time; so these factors must be considered in the high temperature resistance to deformation.

The free energy of a solid metal is a minimum when the solid is a single crystal and all the atoms except those on the surface are located in definite lattice sites. When the number of grains is increased so that the solid becomes a polycrystalline aggregate, fewer atoms are located in lattice sites (some atoms form the grain boundaries) and the free energy of the solid is increased. Thus fine-grained aggregates exist in a higher energy state than coarse-grained aggregates. The inactivity of the atoms prevents grain growth at low temperatures so that the fine-grained structure can persist and offer a maximum resistance to deformation by slip. At high temperatures, however, the atoms become more active and tend to establish themselves in conditions of lower energy. The finest grained structures are the most unstable and consequently are affected

most by temperature. Thus the greater energy associated with the fine-grained aggregates tends to counteract the greater resistance to deformation offered by the increased grain boundary material of the finer grain structure. This theory postulates that at any given temperature there would exist an optimum grain size which would offer the maximum resistance to deformation. There is considerable evidence which supports this contention, but the presentation of this argument is beyond the scope of this paper. The object of this discussion is merely to point out that the thermal activity of the atoms is a factor which need not be considered in low temperature deformation but which is of great importance at elevated temperatures.

The foregoing considerations indicate that an increase in the resistance to deformation at elevated temperatures may be obtained by adjusting the grain size to the optimum. Greater resistance to deformation can also be obtained by the use of solid solution alloys in place of pure metals. The nature of the solute atoms will greatly affect the resistance to deformation at both high and low temperatures, but the effect of the temperature on the solute atoms becomes an important factor. Thus solute atoms of tungsten and molybdenum, their effective action being little affected by temperature, offer a much greater increase in resistance to deformation than the same quantity of elements which are more affected by temperature. However, a great increase in resistance to deformation can be obtained with elements such as nickel and chromium provided they are present in sufficient quantity.

The third method for raising the high as well as the low temperature resistance to deformation involves the presence of a precipitated phase. The size, shape, and dispersion of the precipitated phase greatly affect both the high and the low temperature strength. A critically dispersed precipitate offers the maximum resistance to deformation at temperatures which cause no appreciable agglomeration of the particles. At higher temperatures, the finer particles are more active than coarse ones, and they dissolve and coalesce more readily so that it is not surprising to find that when most steels are tested in creep at 1000 degrees Fahr. the more stable pearlitic structure offers a greater resistance to deformation than the extremely unstable martensitic structure. The resistance to deformation decreases as the structure becomes coarsened, but the stability of the structure becomes greater. The increased stability tends to compen-

sate for the decreased resistance to deformation offered by the coarser structure. The greatest resistance over long periods of time at high temperatures should be obtained with the finest structure which does not change appreciably under the imposed conditions. If coalescence of the dispersed phase does occur, the resistance to de-

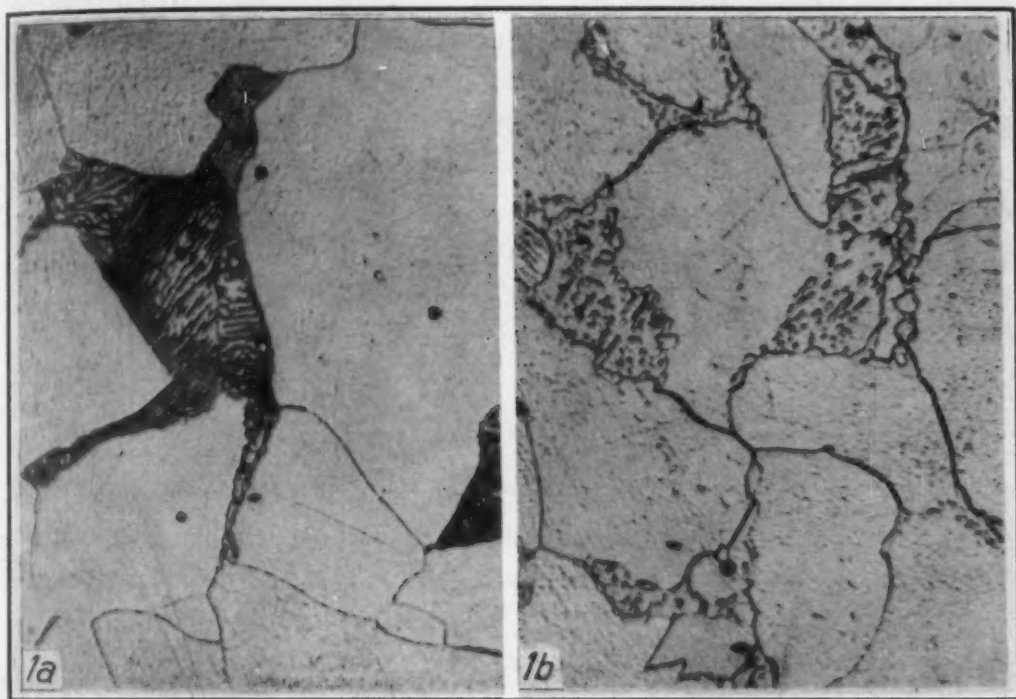


Fig. 1—Change in Structure in Pearlite Area of 0.15 Per Cent Carbon Steel After Long Heating Above 1000 Degrees Fahr.

Fig. 1a—Lamellar Pearlite of Normalized Material.

Fig. 1b—Complete Spheroidization of Fe_3C . $\times 1000$.

formation is correspondingly decreased. Fig. 1a shows a pearlitic structure possessing good resistance to deformation at 1000 per cent Fahr. Prolonged heating at this temperature produces a gradual change in the structure. The pearlite lamellae gradually spheroidize as shown in Fig. 1b. This change in structure produces approximately a tenfold increase in creep rate, which is certainly undesirable. One problem, then, which presents itself is a study of the stability of various structures and the effect of alloy additions on the stability of these structures. Much of this work has been done by competent investigators, and additional work is now in progress. Bailey (2) has shown that an annealed 0.15 per cent carbon steel will spheroidize in 2500 hours at 1000 degrees Fahr., 200 hours at 1100 degrees Fahr., or 20 hours at 1200 degrees Fahr., and that normalizing

or cold working greatly accelerates spheroidization. Bailey (3) also conducted spheroidization tests on carbon-molybdenum steel and found similar results. Alloying elements which form stable carbides greatly increase the time required for spheroidization. The more stable steels available for use at high temperatures require at least three times as long as carbon steel for spheroidization. This increase in stability is very helpful, but much greater stability is highly desirable. Many of the most stable carbides are found in commercial high temperature steels; so it seems desirable to determine the stability and strengthening effect produced by a dispersed intermetallic compound composed of elements other than carbon.

Materials suitable for use under stress at temperatures above 1000 degrees Fahr. should contain a critically dispersed stable phase. Compounds of the type Fe_2W , Fe_3Mo_2 , Fe_3Cb_2 , Fe_2Ta , and Fe_2Ti have been found by the author to be much more stable than the best carbides. High temperature tests showed that alloys containing precipitates of such compounds possessed greater resistance to deformation and fracture at 1100 degrees Fahr. than commercial alloys which depend upon carbides for their strength. An investigation of this nature is necessarily time-consuming so that precise data are acquired slowly. However, it is hoped that the preliminary results reported herein may be of some interest.

EXPERIMENTAL PROCEDURE

The alloys used for testing were melted in an induction furnace. The castings were forged to 1 inch square bars which were heat treated and machined into samples. The room temperature tensile tests were made on the standard 0.505-inch tensile bars. The creep tests were conducted according to the ASTM code for a 12-inch gage length and a 0.505-inch diameter bar. The rupture tests were made with samples two inches long having a 1 inch gage length of 0.1596 inches diameter. The ends of the samples were threaded with quarter-twenty threads. Rupture tests were conducted at 1100 degrees Fahr. by applying various constant loads and measuring the time required for failure in the manner proposed by White, Clark, and Wilson (4). The testing equipment shown in Fig. 2 is designed so that tests can be made in any atmosphere. All tests reported herein were made in air. The test bar, which is suspended in the hot zone of the furnace by long extension rods, is loaded by weights

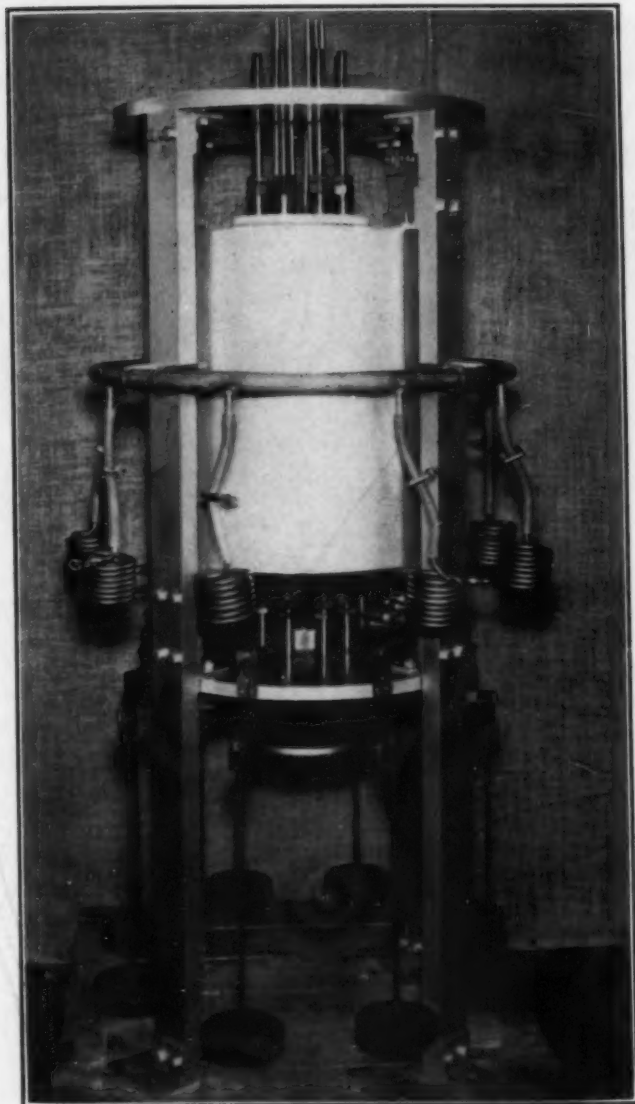


Fig. 2—Testing Equipment for Sustained Load Fractures.

acting through a simple lever. When a bar breaks, a time clock is automatically stopped so that the exact life is recorded.

RESULTS

In addition to the stress-life tests made on all alloys, the recrystallization temperature has been determined for some of the alloys. Since recrystallization is related to stability, it seems worth while to consider briefly some of the results obtained. It was found that the recrystallization temperature was unusually high whenever

a dispersed phase such as Fe_3Mo_2 was present. An example of this is shown in Fig. 3. The structure shown in Fig. 3a was obtained by cold rolling a 3.5 per cent molybdenum iron alloy to 95 per cent reduction, followed by a 24-hour reheat at 1300 degrees Fahr. The solubility of molybdenum in iron is about 3.5 per cent at 1300 de-

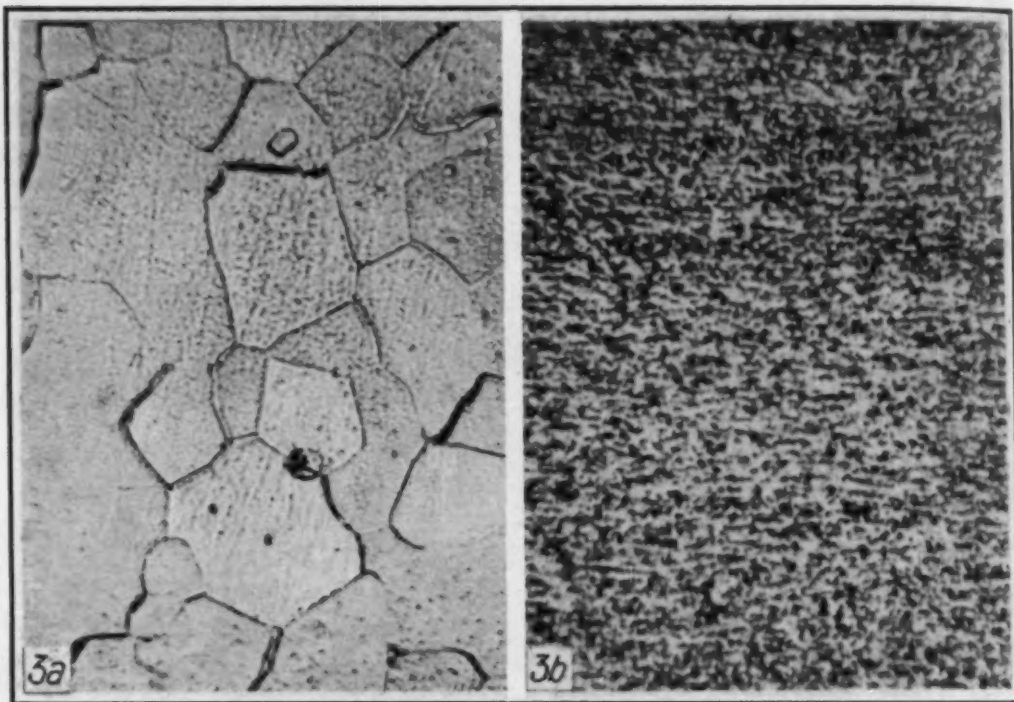


Fig. 3—Iron-Molybdenum Alloys Cold-Rolled to 95 Per Cent Reduction Then Heated to 1300 Degrees Fahr. for 24 Hours.

Fig. 3a—3.5 Per Cent Molybdenum. Fig. 3b—7 Per Cent Molybdenum. $\times 1000$.

grees Fahr.; so this alloy represents the recrystallization of a saturated solid solution. The 7 per cent molybdenum alloy shown in Fig. 3b had exactly the same treatment, but it showed very little evidence of recrystallization. The matrix in each sample was essentially the same, i.e., a saturated solid solution of molybdenum in iron. The higher recrystallization temperature of the 7 per cent molybdenum alloy seems to be due to the blocking effect of the excess molybdenum, present as particles of Fe_3Mo_2 . Similar results were obtained with other alloys which contained dispersed intermetallic compounds. This effect has also been noted in nonferrous alloys (5). The work on recrystallization supported the hypothesis that a dispersed phase of an intermetallic compound which did not agglomerate rapidly at 1100 to 1200 degrees Fahr. had a great stabilizing influence. A heat of the 7 per cent molybdenum alloy was

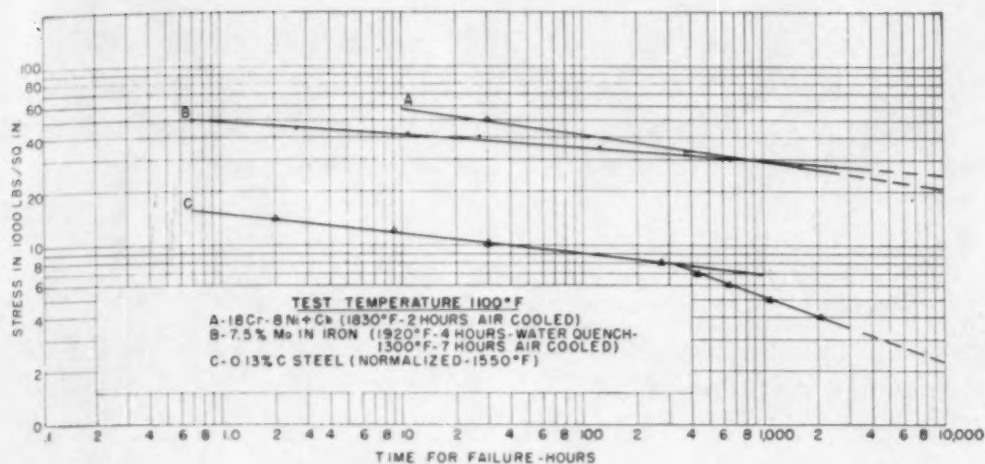


Fig. 4—Curves Showing 1100 Degrees Fahr. Rupture Data of Alloys A, B and C.

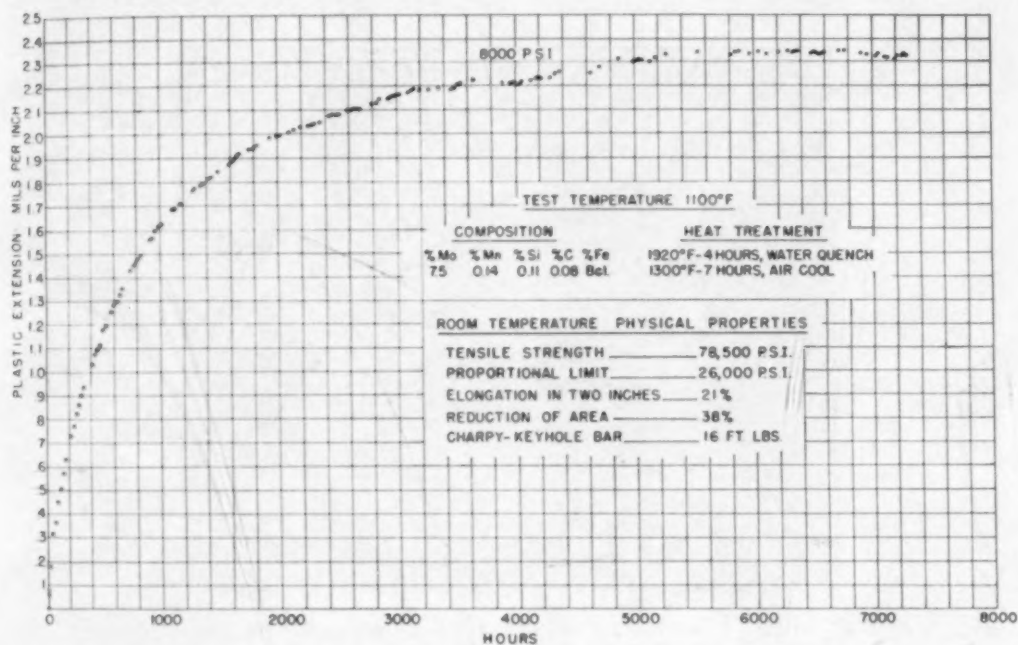


Fig. 5—Creep Data of Alloy B at 8000 Pounds Per Square Inch.

forged, heat treated (1920 degrees Fahr., water quenched; 1300 degrees Fahr., air-cooled), machined and subjected to both creep and rupture tests. The 1100 degrees Fahr. rupture data of this alloy are plotted in Fig. 4 along with the data for columbium stabilized 18-8 and a plain carbon steel for comparison. This alloy was also tested in creep at 1100 degrees Fahr. with a load of 8000 pounds per square inch. The remarkable tendency for this alloy to strain harden at high temperatures is well illustrated by the creep curve shown in Fig. 5. After prolonged loading at this stress and temperature, the creep practically stopped.

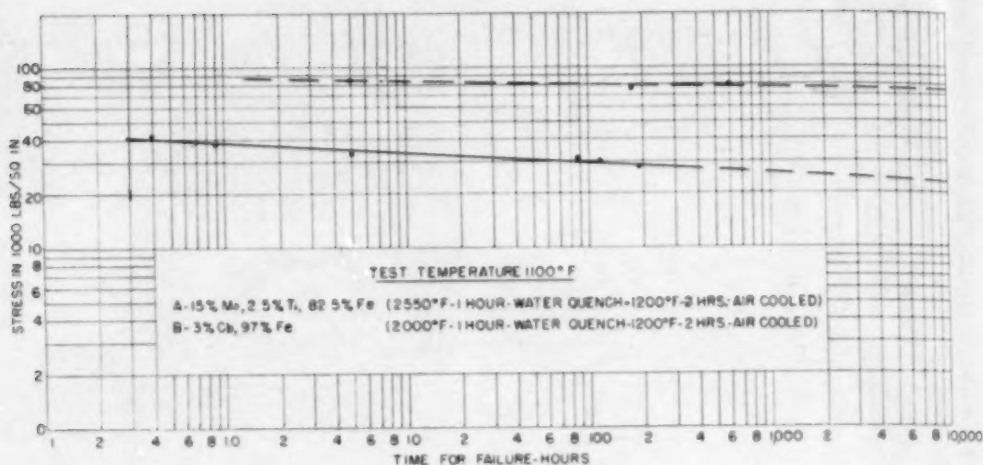
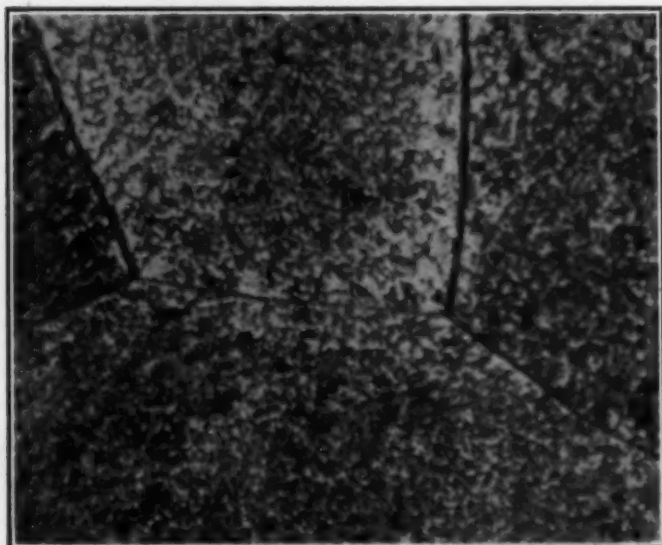


Fig. 6—Rupture Data of Alloys A and B.

Fig. 7—Structure of 3 Per Cent Columbium in Iron After Water Quenching from 2000 Degrees Fahr. and Reheating to 1200 Degrees Fahr. $\times 1500$.

The idea of producing high temperature strength by a finely dispersed stable phase was extended to other alloy systems. Among the more notable alloys are those of iron and columbium. These alloys were cast, forged, heat treated (2000 degrees Fahr., water quench; 1200 degrees Fahr., air cool), and machined as were the other alloys. The data obtained with an alloy containing three per cent columbium, balance iron, are plotted in Fig. 6. The structure of this alloy is shown in Fig. 7. This relatively low alloy has exceptionally good rupture strength at 1100 degrees Fahr. Fig. 6 also includes the incomplete data available for a precipitation hard-

ening alloy of iron containing 15 per cent molybdenum and 2.5 per cent titanium. The data are by no means conclusive, but they do illustrate that the rupture strength of precipitation-hardening alloys can be much greater than steels which depend upon carbides for their strength.

CONCLUSIONS

It is suggested that the factors which affect the low temperature resistance to deformation of an alloy also affect the high temperature resistance to deformation, but the influence of temperature on the various factors must be considered.

It is suggested that an alloy containing a critically dispersed precipitate of a stable intermetallic compound should possess the greatest high temperature strength. Some data are given which verify this suggestion.

ACKNOWLEDGMENTS

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Bibliography

1. E. C. Bain, "Function of The Alloying Elements in Steel," American Society for Metals, 1939, p. 66.
2. R. W. Bailey and A. M. Roberts, "Testing of Materials for Service in High-temperature Steam Plants," IME, Feb. 1932, p. 209.
3. R. W. Bailey, "The Utilization of Creep Test Data in Engineering Design," IME, 1935, p. 140.
4. A. E. White, C. L. Clark, and R. L. Wilson, "Fracture of Carbon Steels at Elevated Temperatures," TRANSACTIONS, American Society for Metals, 1937, Vol. 25, p. 863.
5. R. H. Harrington, "Age Hardening of Metals," American Society for Metals, 1940, p. 314.

DISCUSSION

Written Discussion: By R. F. Miller, research laboratory, United States Steel Corp., Kearny, N. J.

Whereas it is widely recognized that creep occurs in metals under stress at elevated temperature, there is less general appreciation of the fact that the creep strength of a given material depends not only upon its composition but also upon the prior heat treatment, and that exposure to service temperature

for a prolonged period may in itself constitute a heat treatment resulting in change of structure and corresponding change in creep strength. We are glad to see these points emphasized anew by Mr. Parker.

As the author points out, one of the most important considerations in the use of metals at elevated temperature is the determination of the effect of alloying elements and prior heat treatment on structural stability. In his book "Functions of Alloying Elements in Steel", E. C. Bain discusses the difference in behavior and stability between iron carbide and the more slowly diffusing special carbides, which he claims cause "secondary hardening" and improvement in the creep strength of steel. Secondary hardening is considered to be a form of age-hardening, and overaging results from a coarsening of the special carbides beyond the most effective fine particle size.

The improvement in creep strength of carbon-molybdenum steel brought about by the precipitation of a complex carbide or intermetallic compound of optimum particle size was described in a paper by Miller, Campbell, Aborn and Wright entitled "Influence of Heat Treatment on Creep of Carbon-Molybdenum and Chromium-Molybdenum-Silicon Steel", (TRANSACTIONS, American Society for Metals, Vol. 26, 1938, p. 81). In this work, specimens of carbon-molybdenum steel, normalized at 1650 degrees Fahr. and tempered at 1200, 1300, and 1400 degrees Fahr. for both 5 hours and 168 hours respectively, were tested in creep. The tempering operation produced a fine precipitate in the ferrite matrix, and a tempering time of 5 hours at 1300 degrees Fahr. seemed to yield the optimum particle size of precipitate for maximum creep strength and stability; longer tempering at 1300 degrees Fahr. (for 168 hours) resulted in coarsening of the precipitate and lower creep strength.

A short quotation from the above paper may be worthwhile. "The difference in behavior of the pearlitic carbides and the groundmass precipitate suggests a difference in composition. The spheroidization and migration to the grain boundaries and agglomeration of the so-called 'pearlitic' carbides bears a close similarity to the behavior of cementite in straight iron-carbon alloys, and this material is probably an iron-rich molybdenum carbide. On the other hand, the groundmass precipitate forms slowly in fine particles in the ferrite and agglomerates *in situ*. While the composition of this fine precipitate is not definitely known, its appearance and behavior suggests that it is a molybdenum-rich carbide or possibly an iron-molybdenum intermetallic compound. The stability of such a precipitate probably accounts for its lasting ability to improve the creep resistance of the steel, as was suggested by E. C. Bain in 1934.

"The creep rate seems to be associated with the particle size and degree of dispersion of the shower precipitate in the ferrite groundmass.

"The behavior of the precipitate as demonstrated by the changes in microstructure offers a tangible reason for the improvement in creep resistance imparted to steel by the addition of molybdenum."

Study of the improvement in creep strength brought about by the presence of precipitated particles of stable carbides or intermetallic compounds, or both, and of the heat treatment required to form a stable structure and optimum particle size for each service temperature, will undoubtedly assist in the de-

velopment of alloys better suited for use at elevated temperatures. Whether commercial alloys should be entirely carbon-free, or whether the presence of certain carbides in addition to the intermetallic compounds would be beneficial to creep strength and other mechanical properties, is a question on which we would appreciate the author's comments.

Written Discussion: By R. L. Wilson, assistant metallurgist, Climax Molybdenum Co., Canton, Ohio.

The researches reported in this paper are of fundamental importance in studying the problem of producing stronger alloys for service at elevated temperatures. While attention has been directed to the influence of recrystallization temperature on stability of microstructure in alloys heat treated to develop a critically dispersed precipitate, it might be well to examine the same observations from another perspective.

From theoretical considerations, the best way to increase the high temperature strength of metals is to raise the lowest temperature of recrystallization, which is in effect what has happened in these experiments. In general, the conditions that hinder grain growth in metals also tend to raise the lowest temperature of recrystallization after severe cold working. For this reason the iron-molybdenum alloy containing 3.5 per cent molybdenum has a higher recrystallization temperature than the solvent iron, and the iron-molybdenum alloy containing 7 per cent molybdenum has a still higher recrystallization temperature due to the presence of a relatively stable disperse phase of an intermetallic compound.

The increase in high temperature strength in the examples cited in the paper could then be ascribed to raising the recrystallization temperature through the agency of forming a critical dispersion within the alloy, rather than crediting the improvement directly to the stability of the disperse phase. Such a distinction may seem academic but is helpful in explaining the mechanism of strengthening. When an alloy acquires greater resistance to deformation as a result of the precipitation of a disperse phase, the increase in strength originates in the greater resistance to plastic deformation of the individual crystalline grains. However, when the metal is above its lowest temperature of equi-cohesion, the individual grains may be strong but the whole metallic aggregate may be weaker due to low intergranular cohesion.

Consequently, if the lowest temperature of recrystallization is accepted as being the same as the lowest temperature of equi-cohesion for all practical purposes, it follows that the strengthening of alloys for high-temperature service by employing disperse precipitates depends primarily upon the rise in lowest temperature of recrystallization. The creep data shown for Alloy B at 1100 degrees Fahr. under 8000 pounds per square inch would support this viewpoint also. The long-time tensile-rupture tests will provide additional pertinent information, and we shall await with interest the reports on the continuation of this valuable work.

Author's Reply

I should like to extend my thanks to Mr. Wilson and to Dr. Miller for their interest in this paper and for their contributions in the discussion.

Perhaps one thing which should be discussed more fully is the alleged relation between the "equi-cohesive" temperature and the lowest temperature of recrystallization. It is probably safe to say that both of these terms are ambiguous. We have given up the idea of correlating the "equi-cohesive" temperature and the recrystallization temperature because we have found a number of instances in which alloys would recrystallize at temperatures far below those necessary for the formation of intercrystalline fractures. As an example, in high-chromium iron alloys, it is impossible by any method which we have employed or know of to produce intercrystalline failures below 1300 degrees Fahr. and in some cases below 1500 degrees Fahr. Yet all of these alloys will recrystallize in less than a day at temperatures in the neighborhood of 1100 degrees Fahr. Another example was found in the rupture testing of commercial aluminum. This metal would recrystallize from the cold-worked state when heated between 300 and 400 degrees Fahr. But we had to make rupture tests on this material above 700 degrees Fahr. in order to produce an intercrystalline failure. So we no longer associate "equi-cohesive" temperature and recrystallization temperature, even though in many cases they are essentially the same.

Dr. Miller asked whether or not alloys for high-temperature service should be entirely carbon free. It is my opinion that alloys used commercially should possess good properties at both high and low temperatures. It is, of course, desirable to increase the low temperature strength of iron-base alloys with carbon; so it seems that commercial alloys should depend upon carbides for their low temperature strength and upon the more stable intermetallic compounds for high temperature strength.

THE SIGNIFICANCE OF HYDROGEN IN THE METALLURGY OF MALLEABLE CAST IRON

BY H. A. SCHWARTZ, G. M. GUILER AND M. K. BARNETT

Abstract

The paper is in the nature of an exploratory survey of the significance of hydrogen content of white iron intended for conversion into malleable cast iron. Until the development of a reasonably easy method for the quantitative determination of hydrogen, opinions on this subject were largely based on qualitative evidence or even on conjecture.

Data are given on the range of hydrogen contents to be expected in commercial melting and on its relation to oxygen content. A brief exploration of the solubility of hydrogen in graphitized white cast iron is described and it is shown that in the annealing operation hydrogen is largely eliminated.

It is shown that white iron anneals relatively slowly in hydrogen-rich atmospheres and that white cast irons rich in hydrogen initially anneal more slowly than those low in that element. The effect of hydrogen on number of graphite nuclei is recorded.

Malleable cast iron made from metal initially high in hydrogen content is shown to be inferior in tensile properties to metal lower in that element even though the final hydrogen content is quite similar. No mechanism for this impairment has yet suggested itself.

Weight for weight, hydrogen appears the most potent element so far investigated in its effects on graphitizing rate.

INTRODUCTION

IT was the original purpose to prepare a paper dealing with the effect of hydrogen on graphitizing rate. The technique of making additions of metallic elements and determining graphitizing rates is pretty well understood and the results at least fairly reproducible and capable of quantitative evaluation. It soon became evident that a study of the effect of hydrogen is complicated by variables which

A paper presented before the Twenty-second Annual Convention of the Society held in Cleveland, October 21 to 25, 1940. Of the authors, H. A. Schwartz is manager of research, and G. M. Guiler and M. K. Barnett are research chemists, research laboratory, National Malleable and Steel Castings Co., Cleveland. Manuscript received April 26, 1940.

require evaluation before any straightforward graphitization program is possible.

It will presently appear that the execution of graphitization rate measurements in which hydrogen concentration is the only variable is so difficult that an understanding of the subject requires attention to many incidental phenomena which therefore seem to deserve investigation as much, or more, than the narrower field originally contemplated and still far from completely investigated.

SOLUBILITY OF HYDROGEN IN PURE IRON

A diatomic gas which dissolves in the atomic state, as may be expected in the iron-hydrogen system, will have its limiting solubility determined by the degree of dissociation of the gas atmosphere (which is a function of temperature). The partial pressure of the atomic gas will, at a given temperature, be proportional to the square root of the pressure. The solubility of hydrogen in iron, as a function of temperature, is therefore meaningless unless we specify the partial pressure of hydrogen in the atmosphere.

The most commonly accepted data for the solubility of hydrogen in iron are Sieverts', cf. Boyles (1)¹ who worked under a hydrogen pressure of one atmosphere. The dissociation of hydrogen at a total pressure of one atmosphere has been critically examined by Giaque (2) on the basis of energy calculations based on spectroscopic data which should furnish the most accurate basis for conclusions on this point. Sieverts' data and Giaque's are plotted in Fig. 1.

Since the dissociated hydrogen atoms occupy twice the volume of the molecular gas before dissociation, the partial pressure of hydrogen (atoms) is twice the fraction of H_2 dissociated. Assuming the validity of Raoult's law, we may then calculate the vapor pressure of hydrogen at any convenient concentration. The standard concentration is here taken as 1.00 per cent (by weight); this is a purely fictitious value selected as a matter of convenience and it is not implied that a 1.00 per cent hydrogen solution in iron exists or would follow Raoult's law if it did exist.

Fig. 2 is an enlarged section of Fig. 1 in the region from 400 to 1000 degrees Cent. and shows the calculated (a wholly imaginary) vapor pressure of hydrogen from a 1.00 per cent solution in iron as a function of temperature on a logarithmic scale from 400 to 600 de-

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

degrees Cent. and on a linear scale from 500 to 1000 degrees Cent. The data for gamma iron are extrapolated below 900 degrees Cent. on the basis of a rectilinear extrapolation of Sievert's data which is not beyond criticism. One would expect to find the vapor pres-

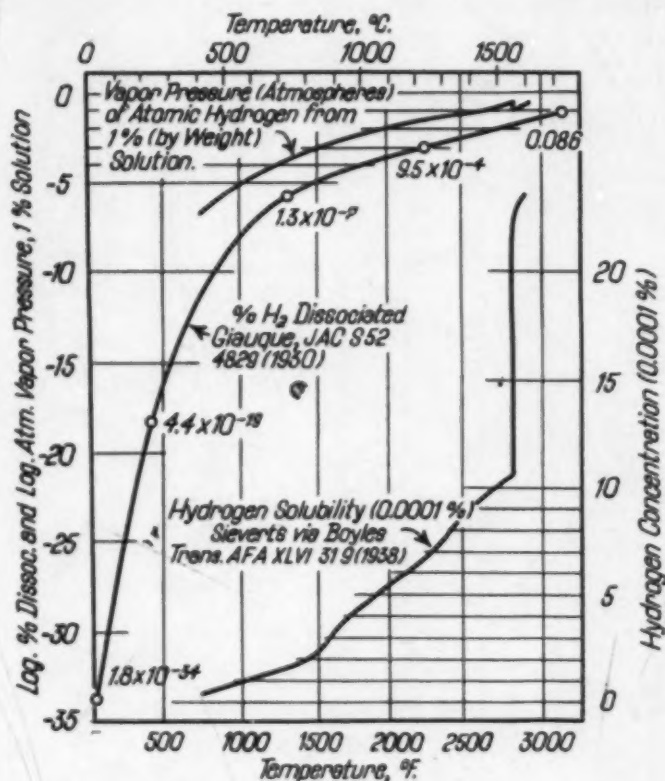


Fig. 1—Dissociation of Hydrogen Compared with its Solubility in Pure Iron.*

sure of hydrogen from a solution of given concentration to follow the Clapeyron-Clausius equation.

In Fig. 3 log. vapor pressure is plotted against the reciprocal of absolute temperature. The points for gamma iron fall well on a straight line as they should and those for alpha iron fall moderately well on another straight line. The upper left hand point being delta iron, should fall on a prolongation of the alpha iron line, but does not do so. Latent heats of vaporization of hydrogen have been calculated from the slopes of the lines and recorded on the diagram.

The lines of Fig. 3 should be a better means of interpolating or

*Sieverts' original publication is unfortunately not before the writer and it is not known whether he measured gas volumes at room temperature or at 0 degrees Cent. Fig. 1 is calculated on the former supposition. If this be incorrect, then the weight per cent of hydrogen plotted in Fig. 1 should be increased in the ratio $\frac{298}{273}$, i.e., by about 10 per cent. The vapor pressure curve must then be dropped by 0.046, a correction not visible on the scale of the figure.

extrapolating in Fig. 1 than that figure itself. Our logarithmic scale is, however, too small to permit of the desired precision especially with regard to Giaque's data. The effect of calculation from two logarithmic curves is to permit inflections to exist such as are found in Sievert's curve even where no allotropic change is involved.

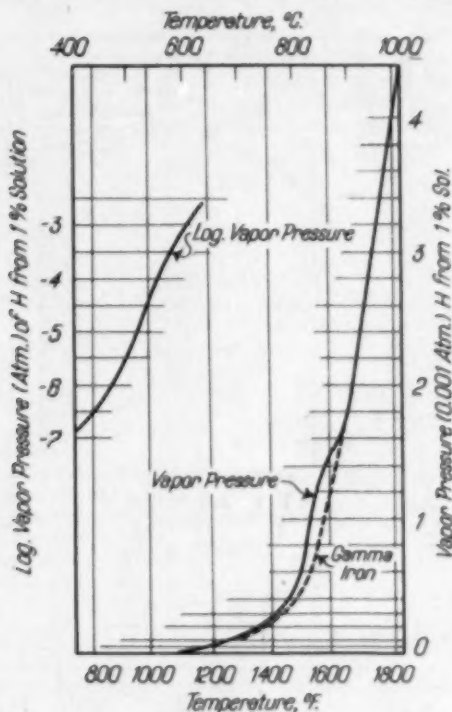


Fig. 2—Vapor Pressure of Hydrogen from a Theoretically 1 Per Cent Solution in Iron.

DETERMINATION OF HYDROGEN IN IRON

The determination of hydrogen by the vacuum fusion method is at best unsatisfactory. Suggestions have been frequently made that hydrogen could be determined by direct combustion in a manner analogous to the ordinary carbon combustion and Schwartz and Guiler (3) have given details of an apparently satisfactory procedure which was used throughout the present investigation.

HYDROGEN SOLUBILITY IN WHITE CAST IRON

An obvious first step in any investigation of the effect of hydrogen is to determine the amount of hydrogen which white cast iron, or the malleablized product made from it, may be expected to retain at various temperatures. White cast iron, for the present purpose,

when referred to without qualification may be considered to mean a material containing approximately 2.5 per cent carbon and 1.0 per cent silicon. Material of this character when used as the cathode in the electrolysis of water at room temperature, in the form of strips about $\frac{1}{8}$ inch thick, current density 0.036 amp. per square inch for 96 hours gains hydrogen from the atomic (nascent) element liberated at its surface till a concentration of about 0.00148 per cent is reached. It is not certain that this is an equilibrium value, but the gain in hydrogen apparently is not very rapid at this time. The value is without meaning, however, for we cannot associate it with any particular partial pressure of atomic hydrogen.

Samples of material so enriched in hydrogen for at least 96 hours were crushed to fragments from 0.05 to 0.08 inch in "diameter", were heated in a very slow stream of tank hydrogen, one bubble in several seconds, at a pressure about 1 centimeter of mercury above barometric in the same tube with similar samples of the unenriched iron which contained about 0.0003 per cent hydrogen. The time of heating varied with the temperature and was so chosen that the two samples finished with approximately equal hydrogen concentrations when equilibrium was considered to be reached. Unfortunately the equilibrium value at 500 degrees Cent. was below the initial value.

The results of this experiment were as follows:

Table I
Solubility of Hydrogen in Solid White Cast Iron after Graphitization

Heat Treatment	Hydrogen Content Normal Iron	Enriched Iron	Probable Approx. Equilibrium Value	Approximate Partial Pressure H (Gaugue)
None	0.00030	0.00148 (+)
443 hrs. at 500° C.	0.00015	0.00018	16×10^{-10}	0.00016
111 hrs. at 700° C.	0.00030	0.00041	20×10^{-7}	0.00030
100 hrs. at 850° C.	0.00037	0.00047	4×10^{-6}	0.00042
Melted with H ₂ at about 1150° C. 4 hrs. frozen in H ₂	0.00150*

*Some hydrogen may have escaped in freezing.

Two facts are to be considered in interpreting the above table. Except the last sample, the iron has certainly been graphitized approximately to equilibrium during hydrogenation and it may have become decarburized. The latter phenomenon is to be expected since it has been observed that a stream of tank hydrogen purified from traces of oxygen or moisture by bubbling through a trap containing the liquid alloy of sodium and potassium continually took up carbon

(probably as CH_4) from white cast iron at temperatures as low as 200 degrees Cent. The stream of hydrogen in the present case did not materially decarburize the iron, being slow though therefore perhaps contaminated with CH_4 . The iron after hydrogenation at 500 degrees Cent. contained only 0.05 per cent carbon less (2.53 per cent) than in its original form.

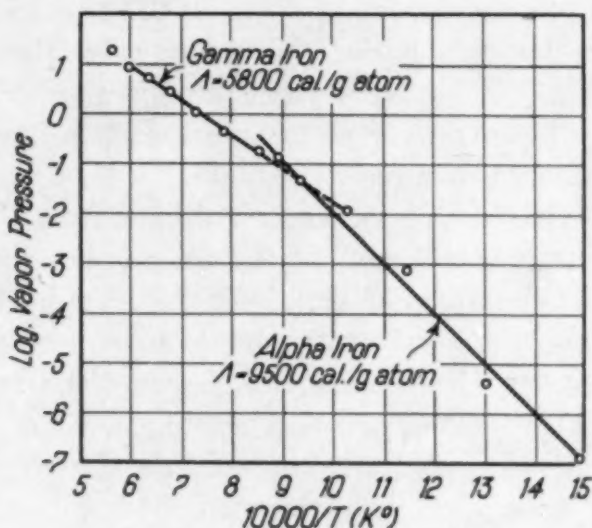


Fig. 3—Vapor Pressure of Hydrogen from a 1 Per Cent Solution in Iron as a Function of Temperature.

The data of Table I are compared with the curve of Sievert's in Fig. 4 in which open circles represent the hydrogen content of graphitized white cast iron, the last value of the table being assigned to solid white cast iron at its (approximate) eutectic melting point. Sievert's curve applies to alpha iron between 700 and 900 degrees Cent., the authors' cast irons are, of course, in the gamma state in that temperature range.

The hydrogen recorded in Table I is no doubt in part dissolved in iron (silico ferrite or silico austenite) and in part held by the graphite.

There was available a silico ferrite containing 0.02 per cent carbon, 0.18 per cent manganese and 0.77 per cent silicon, and approximately 0.125 per cent phosphorus, 0.130 per cent sulphur and 0.06 per cent copper. This material in its initial condition and also electrolytically saturated with hydrogen as described above, was heated 214 hours in hydrogen at atmospheric pressure when, regardless of starting conditions, it contained 0.00042 per cent of hydrogen as compared with 0.00030 per cent. The silico ferrite was lower in

silicon and manganese than the hard iron. One may at least conjecture that silicon increases the solubility of hydrogen in alpha iron and manganese perhaps decreases it. The single experiment here recorded does not suggest that graphite adsorbs much hydrogen. The observation is recorded as a dot in Fig. 4.

HYDROGEN CONTENT OF COMMERCIAL WHITE CAST IRON

Schwartz and Guiler (*loc. cit.*) encountered irons ranging from 0.00018 to 0.0014 per cent hydrogen (weight per cent) and gave brief comments regarding the effect of certain variations in melting process. The matter may well be somewhat expanded here.

Three samples of cupola-electric furnace duplexed malleable were at our disposal of approximately similar, but not identical, composition and produced (a) in the late summer under moderately moist conditions using scrap not abnormally rusty; (b) in the late summer under roughly similar atmospheric conditions but from very rusty scrap; (c) in winter under very dry conditions using scrap presumably similar to (a). The hydrogen content of these three samples was as follows:

Table II
Hydrogen Content of Cupola-Electric White Cast Iron in Relation to Conditions of Charge and Atmosphere

Atmosphere	Scrap	Weight Per Cent Hydrogen
Moist	Normal	0.00083
Moist	Rusty	0.00115
Dry	Normal	0.00033

It is quite apparent that moisture in the air or rust in the charge increases the hydrogen content. In this case we cannot differentiate between a hydrogen pick-up in the cupola and in the electric furnace.

We have available also a series of samples made without special selection of scrap on two wet and two moderately dry days, within an interval of a few weeks, by cupola-air furnace duplexing in which both the cupola and air furnace metal were sampled. The hydrogen data on this material are as follows:

Table III
Hydrogen Content of Cupola-Air Furnace White Cast Iron in Relation to Atmospheric Conditions

Atmosphere	Weight Per Cent Hydrogen		
	Cupola	Air Furnace	Gain
Dry	0.00020	0.00015	-0.00005
Dry	0.00027	0.00055	0.00028
Rainy	0.00070	0.00083	0.00007
Rainy	0.00096	0.00098	0.00002

A series of consecutive analyses on a single dry day, taking samples at roughly one hour intervals from the cupola and air furnace in a duplexing operation is of some interest as throwing light on uniformity of results.

Table IV
Hydrogen Content of Cupola-Air Furnace White Cast Iron on Successive Samples

Time	Weight Per Cent Hydrogen		
	Cupola	Air Furnace	Gain
11:30 A.M.	0.00040	0.00055	0.00015
12:20 P.M.	0.00055	0.00060	0.00005
1:00 P.M.	0.00021	0.00040	0.00019
1:45 P.M.	0.00020	0.00030	0.00010
		Average	0.00012

What variations, if any, of cupola charge occasioned the drop between 12:20 P.M. and 1:00 P.M. is unknown. There is apparently a consistent small gain in the air furnace irrespective of the hydrogen

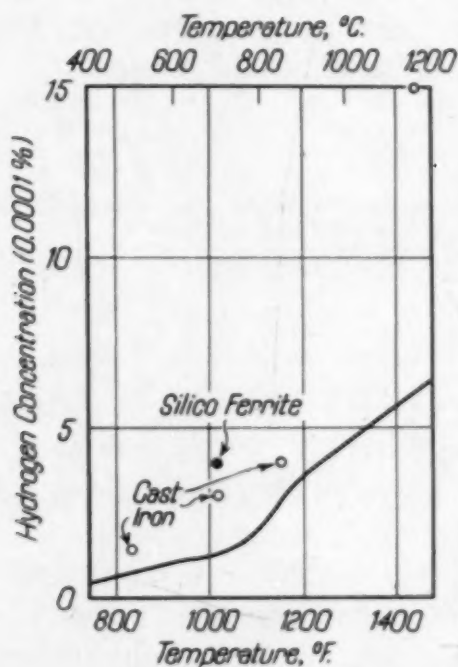


Fig. 4—Effect of Alloying Elements on the Solubility of Hydrogen in Iron.

content of the cupola metal. Reference may also here be made to the statement of Schwartz and Guiler, loc. cit., that "a three day average run using rusty scrap in cupola-air furnace duplexing yielded 0.00062 per cent hydrogen as against 0.00052 per cent hydrogen for clean sprue."

We may also add data on iron made under commercially similar conditions (with one exception, at the same time) as the above in the Brackelsberg furnace.

Table V
Hydrogen Content of Brackelsberg Furnace White Cast Iron in Relation to Atmospheric Conditions

Atmosphere	Weight Per Cent Hydrogen	
	Tap	End of Heat
Dry	0.00030	0.00060
Dry	0.00050	0.00045
Rainy	0.00090	0.00083
Rainy	0.00083	0.00094

The addition of solid metallic alloys to molten white cast iron seems to produce, in general, a marked reduction in hydrogen content. The following data are all on iron melted in the Brackelsberg furnace.

Table VI
Effect of Solid Alloy Additions on Hydrogen Content of White Cast Iron

Alloying Element	Nominal Addition Per Cent	Weight Per Cent Hydrogen		
		Without Addition	With Addition	Loss
Mo	0.23	0.0014	0.0007	0.0007
Mo	0.45	0.0014	0.0005	0.0009
V	0.16	0.0014	0.00011	0.00129
Mn	0.37	0.00019	0.00009	0.0001
Ni	0.25	0.0007	0.00045	0.00025
Ni	0.50	0.0007	0.0005	0.0002
Ni	0.75	0.0007	0.0005	0.0002
Ni	1.00	0.0007	0.00046	0.00024
Cu	0.25	0.0007	0.00055	0.00015
Cu	0.50	0.0007	0.00055	0.00015
Cu	0.75	0.0007	0.00052	0.00018
Cu	1.00	0.0007	0.00051	0.00019

The alloys referred to above have been investigated as to their graphitizing rates (4), (5) in which investigation hydrogen concentration was not recognized as a variable.

One may note from Table VI that the very abnormally high concentration of hydrogen in the heat to which molybdenum and vanadium were added was very much reduced in hydrogen content by any of the additions but notably further by vanadium than by molybdenum. The low hydrogen concentration of the heat to which manganese was added was but little reduced, and to a somewhat higher final value than accompanied the addition of molybdenum to the preceding heat. The effect of additions of either nickel or copper seemed to be largely independent of the amount of the addition. The final hydrogen concentration was of the same order as that produced by molybdenum additions to metal initially much higher in hydrogen

than that used with copper and nickel. It is to be emphasized that when these specimens were first prepared we had neither thought nor knowledge of hydrogen concentrations or else a single liquid metal would have been used for all the alloy additions. The somewhat meager records since obtainable and here recorded seem to point to a tendency of almost any addition to reduce the hydrogen rapidly to roughly 0.0005 per cent (if more was originally present) and to affect some reduction even from lower concentrations. Vanadium perhaps produces a lower hydrogen than the other elements. Perhaps one may say this of vanadium and manganese (the only active *reducing* agents in the list).

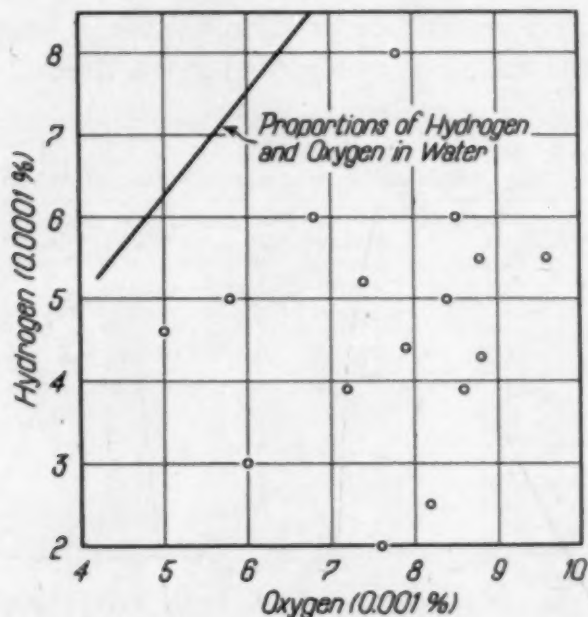


Fig. 5—Oxygen and Hydrogen Contents of Commercial White Iron.

RELATION OF OXYGEN AND HYDROGEN CONTENT OF WHITE CAST IRON

Since rust on the melting stock is shown to increase the hydrogen content of the resulting white cast iron, and since it is also known that rusty stock increases the oxygen content of liquid cast irons, and since the dissociation of moisture in the blast yields both atomic hydrogen and atomic oxygen, one is led to suspect that in any commercial white cast iron some degree of correlation of oxygen and hydrogen content may exist. In Fig. 5 have been plotted the weight percentage of hydrogen in solid metal and of oxygen, in liquid metal,

on a series of samples. The oxygen content is calculated from the Al_2O_3 formed when a large excess of aluminum is added to liquid iron and is probably an accurate determination of the oxygen content of the liquid. It cannot be said with certainty that this amount would survive in the frozen metal had it not been fixed by the addition of aluminum. Since foundry irons do not blow, it seems likely that the comparison in Fig. 5 is reasonably justified.

As might be expected in view of the free oxygen and CO_2 of the flame, the oxygen content of the iron is always more than eight times the hydrogen content as would be the case had the liquid metal merely absorbed the dissociation products of water. It is apparent that there is no close correlation of hydrogen and oxygen content of liquid metal, but some trend toward high oxygen accompanying high hydrogen, and vice versa, can be detected. The slope of such a trend line might not be far from that corresponding to the ratio of the two elements in water.

It must not, however, be assumed from Fig. 5 that, regardless of history and origin, a high oxygen content implies a high hydrogen content. The degree of oxidation incident to the decarburization of basic open-hearth steel is accompanied by a radical decrease in hydrogen content as shown below:

Table VII
Decrease in Hydrogen Concentration with Decarburization of Basic Open-Hearth Steel

	Heat A		Heat B		Heat C	
	Per Cent Carbon	Per Cent Hydrogen	Per Cent Carbon	Per Cent Hydrogen	Per Cent Carbon	Per Cent Hydrogen
Immediately after melting	0.55	0.00070	0.36	0.00080	0.55	0.00090
Before spiegel addition	0.24	0.00022	0.27	0.00060	0.28	0.00070
Before other additions	0.18	0.00004	0.22	0.00040

This is as it should be in view of the FeO-C and FeO-H equilibria.

HYDROGEN CONTENT OF ANNEALED WHITE CAST IRON

From the experiments on the solubility of hydrogen as a function of temperature it is to be anticipated that most white cast iron is supersaturated with hydrogen at partial pressures of hydrogen such as might be expected to exist in heat treating atmospheres where the hydrogen concentration is determined by the rather small dissociation of water vapor either from atmospheric moisture or from the combustion of fuel containing hydrogen.

In Table VIII are recorded the hydrogen contents of white cast irons produced in seven consecutive working days together with the hydrogen contents of the same metal after commercial annealing to malleable cast iron and the loss of hydrogen. The figures are to be interpreted as average values for the days in question.

Table VIII
Hydrogen Content of Unannealed and Annealed White Cast Iron

Weight Per Cent Hydrogen—		Weight Per Cent Hydrogen—	
Hard	Malleable	Hard	Malleable
0.00046	0.000125	0.00085	0.000115
0.00052	0.000085	0.00060	0.000155
0.00050	0.000130	0.00058	0.000140
0.00064	0.000050		
		Average	0.00059
			0.00013

The great decrease to a value approximately independent of the initial hydrogen content is noteworthy.

EFFECT OF HYDROGEN UPON GRAPHITIZING RATE

A really scientific approach to the relation between hydrogen content and graphitizing rate would involve the preparation of two lots of metal precisely alike and differing only in hydrogen content. Each would then have to be graphitized in atmospheres of equal total pressure so adjusted as to partial pressure of hydrogen as to maintain the concentration of that element constant even with changing temperature. Graphitization might then be followed by the usual processes although it would probably not be feasible to treat several specimens together and remove one at a time at predetermined time intervals. We still hope, in the interest of an elegant demonstration, to carry on some such procedure, but for the purpose of the present paper, content ourselves with two much cruder, but perhaps more practical problems: the effect of initially present hydrogen, perhaps complicated by the presence of oxygen, and the effect of hydrogen-rich annealing furnace atmospheres.

Specimens from fifteen heats of very similar composition as to carbon and silicon and of similar origin, which for some reason differed in hydrogen content in the as cast condition, were heated for four hours, under precisely similar conditions, buried in cast iron borings, at 900 degrees Cent. and the resulting graphite determined. It is known that, in principle, graphitizing rate is measured by the amount of graphite so formed (6) although several determinations

on each specimen would make for greater experimental accuracy. The results are as follows:

Table IX
Relation of Hydrogen Content and Graphitizing Rate

Weight Per Cent Hydrogen	Per Cent Graphite Formed in 4 Hrs. at 900 Degrees Cent.	Weight Per Cent Hydrogen	Per Cent Graphite Formed in 4 Hrs. at 900 Degrees Cent.
0.00020	0.69	0.00060	0.515
0.00025	0.54	0.00064	0.028
0.00030	0.42	0.00080	0.23
0.00030	0.40	0.00080	0.185
0.00046	0.38	0.00090	0.080
0.00050	0.085	0.00100	0.15
0.00052	0.018	0.00110	0.08
0.00058	0.037		

It is not perhaps surprising that, with only commercial control on the uniformity of the metal in other respects than hydrogen, there should be a considerable scatter of results, but a definite strong trend downward in graphitizing rate with increasing hydrogen is indicated. Hydrogen from 0.00020 to 0.00046 per cent inclusive averages 0.49 per cent graphite formed in 4 hours. Hydrogen, 0.00050 to 0.00060 per cent inclusive, is very erratic, but averages 0.21 per cent graphite and hydrogen from 0.00064 to 0.00110 per cent averages a little under 0.12 per cent graphite. A decrease in graphitizing rate with increasing hydrogen is evident.

The effect of gaseous environment on graphitizing rate has long been a matter of general interest and random data were accumulated which may now, in part, be correlated with the effect of hydrogen upon the reaction velocity. The matter is best disposed of by citing individual observations. A certain iron held *in vacuo* for 4 hours at 900 degrees Cent. formed 0.155 per cent graphite, but in a stream of hydrogen (stored over water) it formed only 0.075 per cent graphite in the same time. Another sample produced under similar circumstances, 0.31 per cent graphite *in vacuo* and only 0.105 per cent in acetylene, which presumably is completely dissociated at 900 degrees Cent. A third sample producing 0.77 per cent graphite in 4 hours at 900 degrees Cent. in quiescent air, produced only 0.25 per cent graphite in anhydrous NH_3 , which again was presumably completely dissociated. It appears quite evident that atmospheres rich in hydrogen and therefore presumably capable of keeping a relatively high concentration of that element in the metal, seem to have always retarded annealing. In all three experiments the hydrogen content of

the metal may have been continuously variable throughout the heat treatment since attention was then concentrated on the environment without means of correlating the observations with the composition of the specimen.

Though realizing that the technique was not adapted to securing quantitative control of hydrogen concentration, an experiment was undertaken to determine the difference in graphitizing rate for a given iron with a very high and a very low hydrogen content.

A given white cast iron was approximately saturated with hydrogen by being made the cathode (0.05 amperes per square inch) in the electrolysis of a 10 per cent Na OH solution for 48 hours. This is 117,500 coulombs per square centimeter, corresponding to a transfer of 12 gram atoms of a univalent substance, i.e. 12 grams per square centimeter of hydrogen (atomic weight = 1). The specimen was a rod about 1 centimeter square and hence having an area of about 4 square centimeter per cubic centimeter. The amount of hydrogen delivered was therefore about 48 grams per cubic centimeter of metal or about 6 grams of hydrogen per gram of iron. Most of this, of course, escaped into the air as H_2 .

Another portion of the same iron was heated for 620 hours at 420 degrees Cent. in a stream of dry ("pure" tank) nitrogen to remove as much hydrogen as possible. Any fear that material graphitization would take place proved unwarranted, for the metal at the end of the treatment contained but 0.03 per cent graphite.

A portion of the hydrogen impregnated metal was heated in hydrogen at 1 atmosphere for $5\frac{1}{2}$ hours and portions of the initial metal and that which had been treated to reduce the hydrogen so far as convenient for the same time in nitrogen at atmospheric pressure. The graphite at the end of the treatment was then determined with the following results:

Table X
Graphite After $5\frac{1}{2}$ Hours at 900 Degrees Cent. in Relation to Hydrogen Content

Metal Condition	Hydrogen Impregnated Hydrogen	Original Nitrogen	Hydrogen Diminished Nitrogen
Atmosphere	0.05	0.32	1.68
Per Cent Graphite	0.00152	0.00040	0.00009
Wt. Per Cent H Initial	0.00056	Probably decreased	Probably unaltered

The figure 1.68 per cent graphite in the last column represents practically the attainment of equilibrium for the metal whose total carbon

was about 2.55 per cent. No significant decarburization was encountered in any specimen. The correction of the value 1.68 per cent for the 0.03 per cent graphite formed in the preliminary heat treatment is only of academic interest and was omitted. The decreased graphitizing rate accompanying increased hydrogen concentration in a given metal is plainly indicated. Some increase in nodule number as a result of the preliminary treatment in nitrogen was noted.

HYDROGEN'S EFFECT ON NODULE NUMBER

When an element affects graphitizing rate it is always important to determine whether this is accomplished by a change in nodule number. The number of graphite nodules (visible at 100 magnifica-

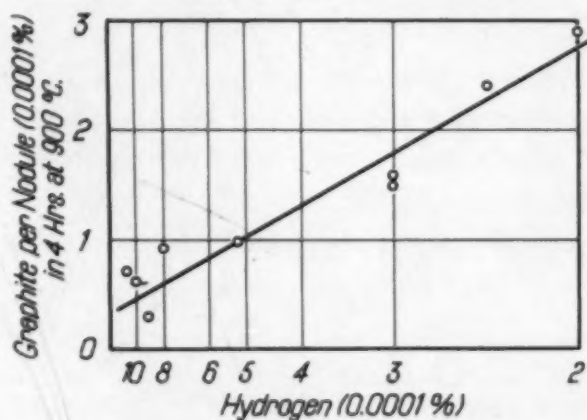


Fig. 6—Growth Rate of Graphite Nodules as a Function of Hydrogen Content.

tion) per square millimeter of actual specimen surface was counted on each of nine specimens of known hydrogen content and relative graphitizing rate. This number was converted into a number per cubic millimeter by methods described elsewhere (7). The results appear below:

Table XI
Nodule Number of Materials Differing in Hydrogen Content and Graphitizing Rate

Weight Per Cent H ₂	Nodules Per Cubic Millimeter	Per Cent Graphite in 4 Hrs. at 900° C.	Per Cent Graphite in 4 Hrs. at 900°C. Per Nodule
0.00020	2350	0.69	0.00029
0.00025	2250	0.54	0.00024
0.00030	2800	0.42	0.00015
0.00030	2500	0.40	0.00016
0.00052	1750	0.18	0.00010
0.00080	2400	0.23	0.000095
0.00090	2600	0.08	0.000031
0.00100	2300	0.15	0.000065
0.00110	1100	0.08	0.000072

The nodule number bears no close relationship to hydrogen content but the rate of growth per nodule does diminish rapidly and rather consistently with increasing hydrogen content. A mechanism in which hydrogen retards the migratory rate of carbon (or less probably, the rate of dissociation of cementite) seems indicated.

In Fig. 6 the reciprocal of the hydrogen content is plotted against the graphite formed per nodule in a given time and an approximately linear relation is seen. An expectation that the logarithm of the graphite per nodule might be a linear function of hydrogen content was not confirmed.

HYDROGEN'S EFFECT ON MECHANICAL PROPERTIES OF MALLEABLE CAST IRON

Even though we have seen that annealed malleable cast iron contains a small and rather constant amount of hydrogen, there seems to be a fairly definite relation between the hydrogen content of white cast iron and the properties of the resulting malleable. Certain data on air furnace duplexed malleable compiled from plant records (except as to hydrogen content) are compiled below.

Table XII
Tensile Properties of Malleable Cast Iron in Relation to Hydrogen Content

Wt. Per Cent Hydrogen	Approximate Per Cent Carbon	Tensile Strength Pounds Per Square Inch	Elongation Per Cent in 2 Inches
0.00020	2.47	52769	18.0
0.00025	2.47	53561	20.0
0.00030	2.47	53488	17.0
0.00030	2.65	44270	6.5
0.00046	2.59	52216	16.0
0.00050	2.59	51987	16.5
0.00052	2.59	52834	14.0
0.00058	2.61	44727	9.5
0.00060	2.61	41532	7.5
0.00064	2.61	45151	7.0
0.00080	2.61	45998	6.5
0.00080	2.65	42575	9.0
0.00090	2.65	45477	11.0
0.00100	2.65	45314	11.0
0.00110	2.65	45216	13.5

Some depreciation of quality might be ascribed to the increase in total carbon, the difference between 2.47 and 2.65 per cent might represent a decrease of about 1500 pounds per square inch in tensile strength and 2 per cent in elongation. Primary graphite is known not to have been present in these test specimens. It is evident that

hydrogen very definitely impairs the properties of the resulting malleable even though the effect is to some extent erratic. The mechanism of this effect is interesting and deserves further study.

CONCLUSIONS

The investigation here reported may be regarded as extensive rather than intensive. Recognizing that we have followed none of the investigations to completion, we still hope that we have substituted quantitative data for conjecture as a basis for at least qualitative conclusion in a field where speculation reigned supreme.

The following applies to hydrogen as determined by direct combustion and more specifically to metal melted by the cupola, coal fired air furnace duplex process. It applies also only to white cast irons of such composition as are commonly encountered in malleable practice since we have not extended these investigations beyond that range.

1. White cast iron (in a partially graphitized state) will dissolve approximately twice as much hydrogen from a given atmosphere as will pure iron at the same temperature according to Sievert's data.

2. On a very meager experimental basis there is no evidence of considerable hydrogen adsorption on graphite.

3. White cast iron made under commercial circumstances may contain from about 0.0002 to 0.0010 per cent or more hydrogen by weight.

4. This hydrogen originates from rust in the charge and from moisture in the air and doubtless from other sources, as fuel, also.

5. Oxidizing reactions during melting reduce the hydrogen content, but in general, the oxygen content of metal high in hydrogen is likely to be high and vice versa though not in any quantitative relationship.

6. The introduction of a small amount of almost any solid metallic addition into liquid white cast iron reduces the hydrogen content.

7. On commercial annealing the hydrogen approaches a low constant value somewhat above 0.0001 regardless of initial content.

8. At room temperature white cast iron will absorb up to about 0.00150 weight per cent of hydrogen when used as the cathode in the electrolysis of water.

9. At temperatures around 400 degrees Cent. or over white cast iron loses its hydrogen content to a hydrogen free atmosphere slowly but rather completely.*

10. Hydrogen from any source very greatly retards the graphitizing reaction.

11. Hydrogen does not seem to affect the "nodule number" of the resulting malleable cast iron but does seem to affect the migratory rate of carbon in solid solution at least in gamma iron.

12. White iron high in hydrogen content as the result of melting practice, produces iron of lower mechanical properties than iron lower in that element.

As a general statement it may be said that hydrogen is, weight for weight, apparently the most potent element normally occurring in white cast iron in its influence on the properties. It is even possible that this statement is true atom for atom. Much work remains to be done to elucidate completely the subjects discussed in this paper and a great expansion of the field of investigation is also desirable. For example, the twelfth conclusion above is at present too far simplified. Additional work is already in progress to determine the mechanism by which an element can affect the mechanical properties of a metal after that element has been removed and to develop methods for the control of that mechanism.

ACKNOWLEDGMENTS

This paper has benefited so much from the cooperation of the entire organization of the National Malleable and Steel Castings Company and others that it becomes difficult to enumerate all the contributors. Our thanks are due to C. H. Junge, W. K. Bock and George Watson of the Research Department of the National Malleable and Steel Castings Company, as well as to Harold H. Johnson, formerly of that department, and to Joseph Spretnak, the senior author's graduate assistant at Case School of Applied Science, for experimental data.

*More recent observations extend the temperature range at which hydrogen is readily removable to much lower temperatures.

Carl Joseph of the Saginaw Malleable unit of General Motors Corporation kindly contributed the specimens of electric furnace metal and Milton H. Tilley, B. C. Yearley, William Zeunik and Harold H. Johnson collected material at the Cleveland, Chicago, Indianapolis, and Sharon Works of the Company.

Bibliography

1. Alfred Boyles, "The Formation of Graphite in Gray Iron," *Transactions, American Foundrymen's Association*, Vol. XLVI, 1938, Fig. 27, p. 319.
2. W. F. Giauque, "The Calculation of Free Energy from Spectroscopic Data," *Journal, American Chemical Society*, Vol. 52, 1930, p. 4829.
3. H. A. Schwartz and G. M. Guiler, "Hydrogen in Solid White Cast Iron," *Transactions, American Foundrymen's Association*, Vol. XLVII, 1939, p. 742.
4. H. A. Schwartz, H. H. Johnson and C. H. Junge, "Retarding Effect of Certain Metallic Elements on Graphitization," *TRANSACTIONS, American Society for Metals*, Vol. 25, 1937, p. 609.
5. H. A. Schwartz, V. Fiordalis, J. L. Fisher, J. F. Shumar and M. J. Trinter, "The Accelerating Effect of Certain Metallic Elements on Graphitization," *TRANSACTIONS, American Society for Metals*, Vol. 28, 1940, p. 143.
6. H. A. Schwartz, "Graphitization at Constant Temperature," *TRANSACTIONS, American Society for Steel Treating*, Vol. 10, 1926, p. 53.
7. H. A. Schwartz, "A Simple Approximate Method for determining Nodule Number," *Metals and Alloys*, Vol. VII, 1936, p. 278.

DISCUSSION

Written Discussion: By A. B. Kinzel, chief metallurgist, Union Carbide and Carbon Research Laboratories, Inc., New York.

The authors are certainly to be congratulated on a searching type of investigation, truly worthy of the name research. The work sheds light on some of the anomalous phenomena encountered in the manufacture of malleable cast iron. The effect of hydrogen on malleableizing rate is interesting, as the authors point out that the nodule number bears no close relationship to hydrogen content but the rate of growth per nodule diminishes rapidly.

The authors state in a footnote that the temperature range at which hydrogen is readily removed has been extended to much lower temperatures by more recent observations. In the work being carried out by Crafts and Burgess at the Union Carbide and Carbon Research Laboratories with which the authors are familiar occasion was had to check hydrogen loss of white iron by heat treatment well below the 400 degrees Cent. cited. It was found that the treatment removed 2.6 cubic centimeters of hydrogen in the 300-gram sample. This may be of interest in view of the authors' conclusion contained in the abstract, that hydrogen appears to be the most potent element so far investigated on the effect of graphitizing rate.

It is a pleasure to confirm the broad effect of heating on hydrogen elimination, particularly at the lower temperatures, and to again congratulate the authors on the clear conception and experimental technique with which this was investigated.

Oral Discussion

W. R. HAM:¹ I am delighted to see the interest that metallurgists are now taking in the importance of presence of hydrogen as well as oxygen in metals, although in studying recently some of the reports in Smithell's book on "Gases of Metals", I ran across one compilation of data that gives results that are abnormal on the solubility of hydrogen in certain metals and the abnormality cannot easily be explained except by the introduction of oxygen in the process of charging these metals electrolytically. I should like to ask the question whether in the process of using the electrolytic method of charging any metal with hydrogen, if at the same time there was not the possibility of introducing oxygen.

Probably in my earlier work I did not have an ordinary solution of hydrogen but a combination of the solution of hydrogen in the metal as well as formation of H_2O . Quite often any abnormality of hydrogen diffusion isotherms may be explained by the presence of oxygen, or in the case of iron and nickel by carbon in solution.

It is interesting to note that the enormous pressures that can be obtained in certain regions within metals by putting gases in solution, is now well recognized. I think putting the samples in a lead bath furnishes a lot better vacuum than you would ordinarily get by any vacuum system.

NEIL METCALF:² May I ask a question? Would Dr. Schwartz like to say what he considers to be the chief sources of hydrogen and what steps are taken to keep it at a minimum.

JOHN CHIPMAN:³ I would like to ask Dr. Schwartz if he has a dependable method for the analysis of hydrogen in the material which will distinguish with certainty between hydrogen and water in iron.

B. S. NORRIS:⁴ I would like to give you an example of how the furnace atmosphere may effect graphitization. I have had the occasion to study graphite formation in some specially prepared iron-carbon alloys containing 1.1 per cent carbon. These alloys were melted and cast in a hydrogen atmosphere. The ingots were then forged and rolled with all heating being conducted in a reducing atmosphere to minimize oxidation.

After an alloy containing 0.07 per cent aluminum was hardened in a vacuum and subsequently tempered in a lead bath, covered with "lead-pot" carbon, for 125 hours at 630 degrees Cent., practically no graphite was present. However, when other hardened samples of this same material were tempered in a vacuum instead of in lead, almost complete graphitization was effected. The

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²Metallurgist, Burlington Steel Co., Hamilton, Ont.

³Department of metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.

⁴Research metallurgist, United States Pipe & Foundry Co., Burlington, N. J.

pressure in the furnace during the vacuum treatment was approximately one-tenth of a millimeter of mercury.

Although the above mentioned graphitization occurred below the eutectoid temperature it is quite possible that the graphitization reaction above the eutectoid also might be influenced by the furnace atmosphere.

It would be interesting to know in connection with Table I the relative effectiveness of the various temperatures on the removal of the hydrogen from the enriched iron. Since the time values listed for 500, 700, and 850 degrees Cent. are different for each temperature, it is impossible to make this comparison with the data listed.

Authors' Reply

Answering Mr. Metcalf we know certainly of two sources of hydrogen, one of which is moisture in the blast, the other of which is rust on scrap, which represents the presence of a decomposed hydroxide and it is quite possible to trace variations of hydrogen content to the presence or absence of rust on scrap and/or the presence or absence to a considerable amount of moisture in the blast. I have no doubt that hydrogen in melting stock in some form other than rust would be equally effective, and I have a reasonably strong opinion that high hydrogen fuels have something to do with this matter.

Now having thought about where it comes from, what you do about it is quite obvious. What you do about it is not to use too rusty a scrap and what you may sometime do is that you may dry the blast that is going over the metal.

Answering Dr. Chipman we would get by the method which we use, which is a direct combustion method, hydrogen in any form that the iron contained it, so that if we had, as Dr. Ham suggested, water or something of that kind, the hydrogen of that water would appear in our analysis. We are quite sure we have not occluded moisture on the surface because of the fact our samples are kept at a fairly high temperature, a temperature well above the boiling point of water, until they are just ready to handle and kept in dry atmosphere. By this method we get all the hydrogen whether in compound or not.

Answering Dr. Norris an explanation of the reason for the variation of time with temperature is found in the paragraph preceding the table containing the data to which he refers. We desired a reasonable approach to equilibrium in each case. The rate of removal of hydrogen as a function of temperature has received some consideration since the paper was written and proves to be an interesting study requiring rather extensive experimentation.

It remains only to thank Dr. Ham for his comments and Dr. Kinzel for his kind expressions with regard to our work and for the additional data on hydrogen removal.

DILATOMETRIC STUDIES IN THE TRANSFORMATION OF AUSTENITE IN A MOLYBDENUM CAST IRON

BY DONALD B. OAKLEY AND JOSEPH F. OESTERLE

Abstract

The paper presents the results of an investigation of the transformation of austenite as found in a 0.50 per cent molybdenum cast iron, and on the base iron from which the alloy was made, at seven constant subcritical temperatures.

A modified dilatometer was used in this study. The total reaction times of many specimens at each temperature was observed. Metallographic pictures of completed reactions are shown.

THE authors are presenting this study as a further contribution to the knowledge of the transformation of various austenites. They were thoroughly impressed by the tediousness of the techniques already developed and the values these methods of study offered to the investigator.

The idea of simplifying reaction-rate studies was aimed toward the production of completed studies in as short a time as possible. This would immediately show the possibility of commercial application, together with the permanence of any commercially produced structures. While it is admitted that much is lost when the progress of transformation cannot be followed and studied metallographically, but at the same time, a dilation report tells a very important story.

SPECIMENS

The test materials used for this work were two types of gray cast iron arbitration bars. These analyzed 2.9 per cent carbon, and 2.1 per cent silicon, and differed only in that one type contained 0.5 per cent molybdenum. These irons were designated numbers 1 and 2, for the base iron and molybdenum iron, respectively.

The specimens were machined to a tubular form (2.000 inches long by 0.375 inch i.d. by 0.500 inch o.d.) from the bars as cast.

A paper presented before the Twenty-second Annual Convention of the Society held in Cleveland, October 21 to 25, 1940. Of the authors, D. B. Oakley is associated with the Bethlehem Steel Co., Bethlehem, Pa., and J. F. Oesterle is associate professor, metallurgical engineering, University of Wisconsin, Madison, Wis. Manuscript received July 1, 1940.

APPARATUS

Dilatometer—The extremely low expansion of fused silica [0.55×10^{-6} per degree Cent. (21)]¹ and its high resistance to heat corrosion warrants it for use in dilation measurements. Thereupon, a dilatometer of this material was developed and used during this work. This instrument is mechanically similar to a quartz dilatometer used by the National Bureau of Standards. It combines the virtues of low expansion in the system, simplicity of construction, rapid quenching action, positive indication of expansion, and high sensitivity, all in a compact unit. A specimen may be mounted in the instrument and run through a complete heat treating cycle, both high and low temperatures, without disturbing it.

The essential parts of the instrument can be noted in Fig. 1. The specimen (a) is fitted over a Vitreosil tube (b) (scant $\frac{3}{8}$ inch o.d. by 18 to 24 inches) and is supported by the fused silica button (c) welded to the bottom of the tube. On top of the specimen, and sliding on the tube, is placed a second Vitreosil tube (d), ($\frac{3}{8}$ inch i.d. by $\frac{5}{8}$ inch o.d. by 12 to 18 inches long). Over the top of the larger tube (d) is a flanged steel cap (e) which fits loosely around the smaller tube (b) at its flange, and helps to hold the outer tube in alignment with tube (b). Over tube (b) the flange was cut to paper-thinness, to eliminate, as much as possible, errors from expansion in the steel. Three screws (f) set loosely around the bottom of the cap, helped to hold it in alignment with the sides of its contained tube as it transferred length changes in the specimen to the Ames dial gage (g) mounted on tube (b). This gage was divided in 0.001 millimeters.

The ends of the larger tube (d), and the top of the button (c) were ground smooth, and at right angles to the tube axes. While in operation the steel cap and Ames gage were cooled by means of an air jet to further reduce errors from their thermal expansion.

Furnaces—For high temperature heating the dilatometer was placed in a Micromax-controlled, vertical tube, electric furnace. Excessive oxidation of the specimen was prevented by heating in an atmosphere of nitrogen.

The constant temperature bath was "low temperature drawing salt," about two gallons in volume, in an electric furnace with auto-

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

matic temperature control. Temperatures were kept constant to within 5 degrees Fahr. of the nominal operating temperature. Temperature gradients across the bath were so small, and the volume of salt so relatively large, that no mechanical stirring arrangement was required.

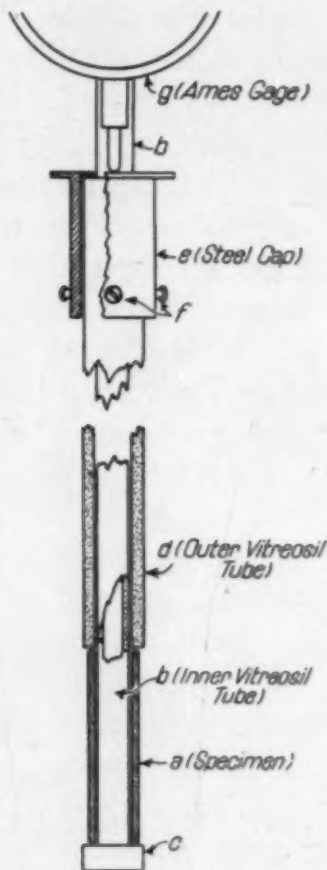


Fig. 1—Oosterle Dilatometer Tubular Specimen, Quartz System, Expansion Gage.

Recording Data—During the relatively short runs data were recorded from visual observation of the Ames gage readings and a stop watch. For long periods an 8-millimeter movie camera was used, photographing at set intervals both the time and Ames gage readings. The camera timing device was operated by a synchronous electric motor, and could be set to operate the camera momentarily at desired intervals, ranging from five minutes to one hour.

PROCEDURE

The general procedure of the research was to run dilations on



Fig. 2—Dilatometer in Preheating Furnace at 1525 Degrees Fahr.

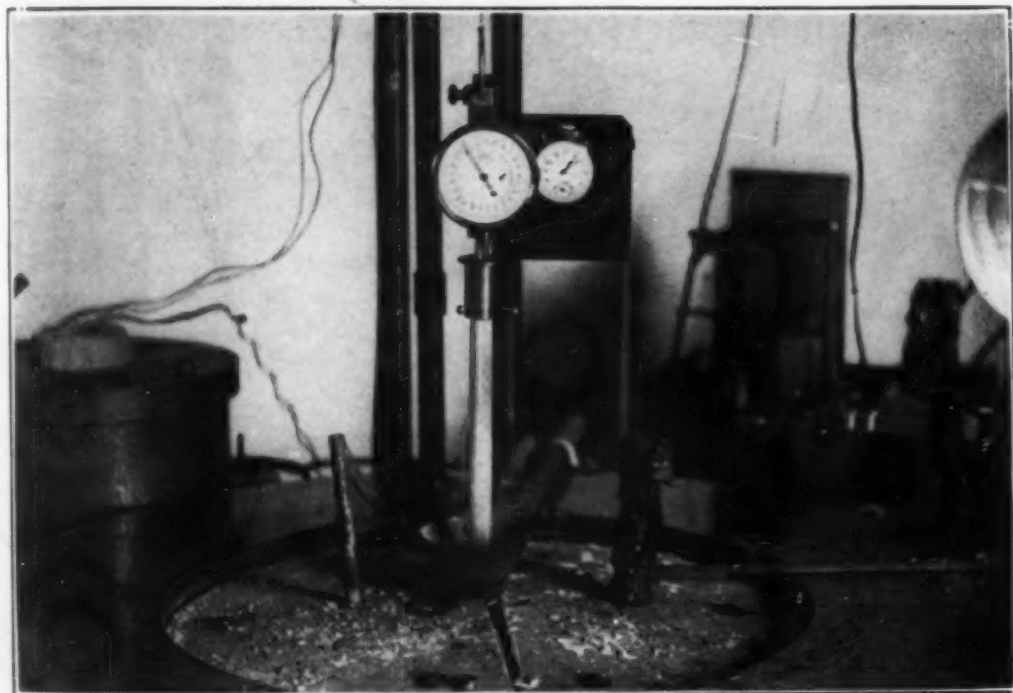


Fig. 3—Dilatometer in Constant Temperature Salt Bath.

each type of iron at 100-degree Fahr. intervals from 1000 to 400 degrees Fahr., inclusive. Dilations were run at each temperature until reasonable checks were obtained. Average values were then

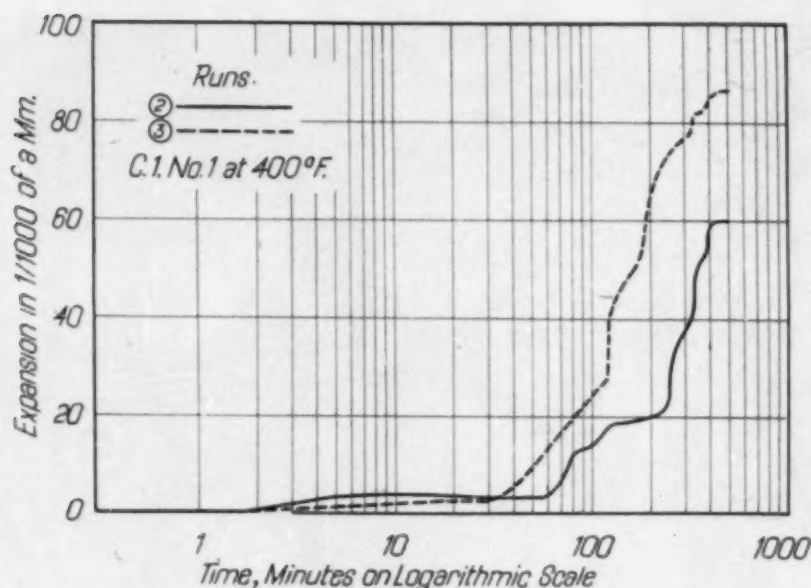


Fig. 4—Expansion-Time Curve of Base Iron Quenched Into Constant Temperature Bath at 400 Degrees Fahr.

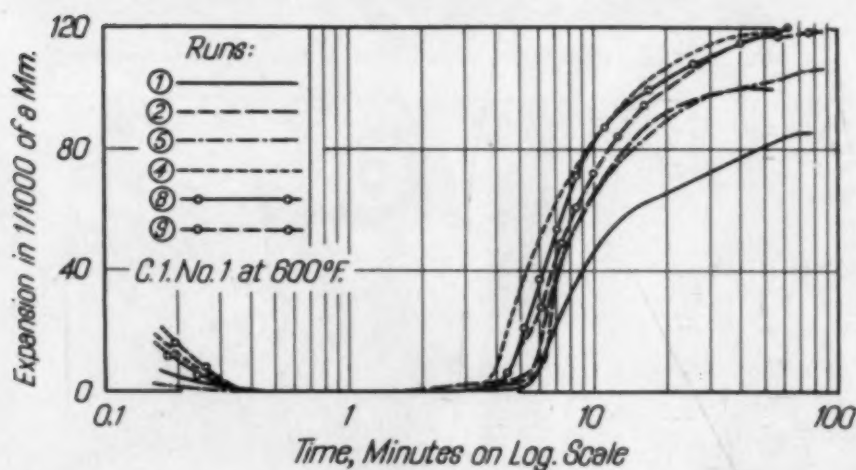


Fig. 5—Expansion-Time Curve of Base Iron Quenched Into Constant Temperature Bath at 600 Degrees Fahr.

taken from these expansion curves for the apparent beginnings and endings of reaction, and plotted on temperature versus log. time axes, to give S-curves.

The high temperature furnace was operated at 1525 degrees Fahr. (830 degrees Cent.). This was purposely made low to help avoid complications from graphite solution. The dilatometer, with a specimen, was kept in the high temperature until the needle of the Ames gage no longer moved. This was usually 15 minutes. At the end of this period the instrument was lifted from this furnace, and plunged quickly into the salt bath. Time and gage readings were

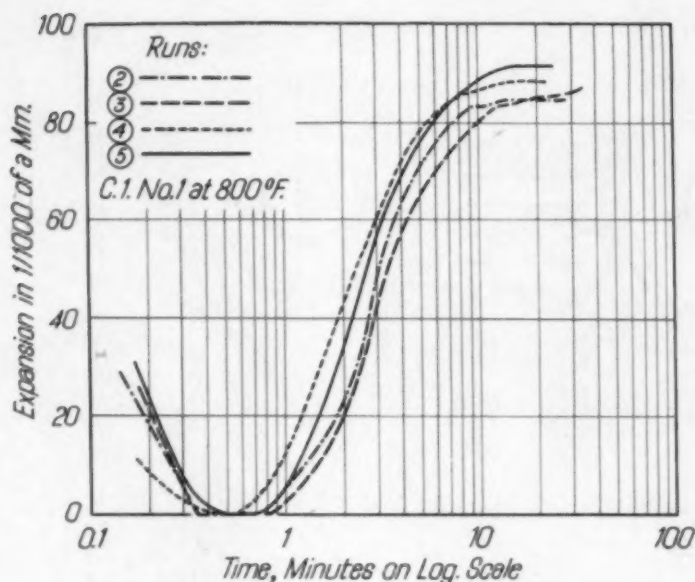


Fig. 6—Expansion-Time Curve of Base Iron Quenched Into Constant Temperature Bath at 800 Degrees Fahr.

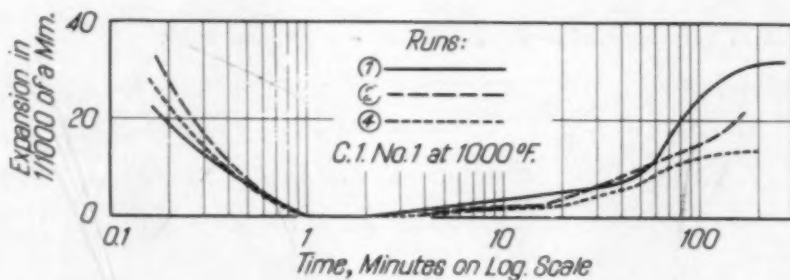


Fig. 7—Expansion-Time Curve of Base Iron Quenched Into Constant Temperature Bath at 1000 Degrees Fahr.

then taken as often as necessary until it was evident that the reaction had completed.

EXPERIMENTAL RESULTS

The accompanying curves (Figs. 4 to 11) represent the expansions of the cast iron specimens as they were quenched to and held at constant subcritical temperatures in the salt bath. Plotting time on a logarithmic scale shows detail at the beginning of expansion, and saves the space that would be necessary if a linear scale were used.

The curves, Figs. 12 and 13, summarize the average times required for reaction at the temperatures studied, and allow a more ready comparison between the reaction times for the two irons.

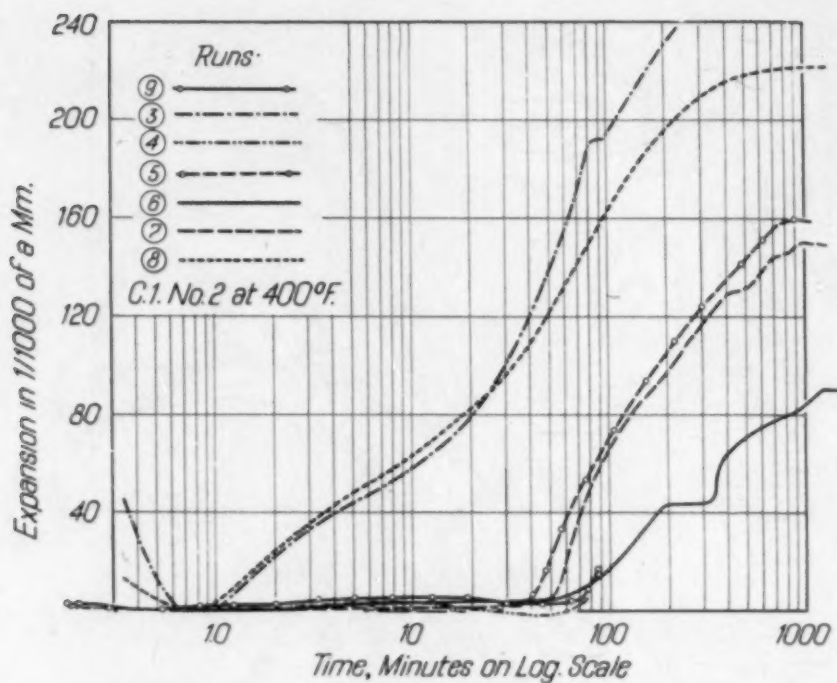


Fig. 8—Expansion-Time Curve of Molybdenum Iron Quenched Into Constant Temperature Bath at 400 Degrees Fahr.

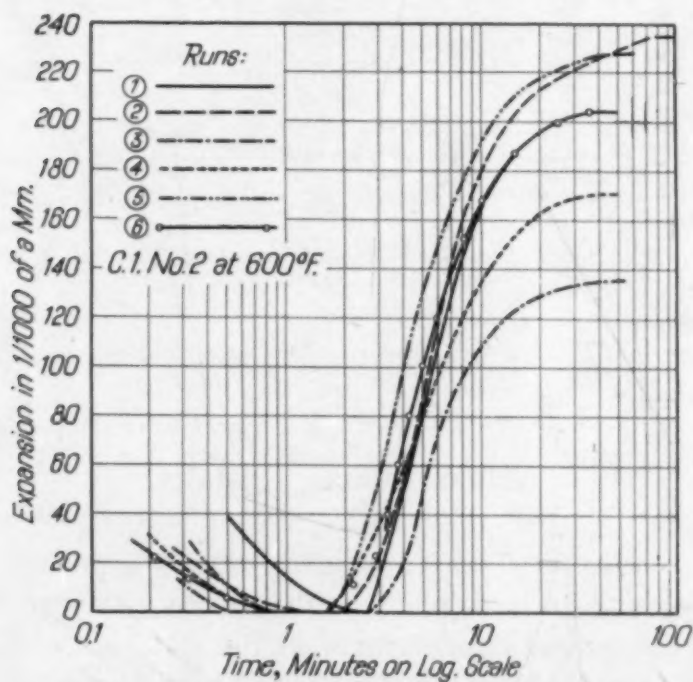


Fig. 9—Expansion-Time Curve of Molybdenum Iron Quenched Into Constant Temperature Bath at 600 Degrees Fahr.

DISCUSSION OF THE CURVES

Although the curves show some irregularities, there can be no

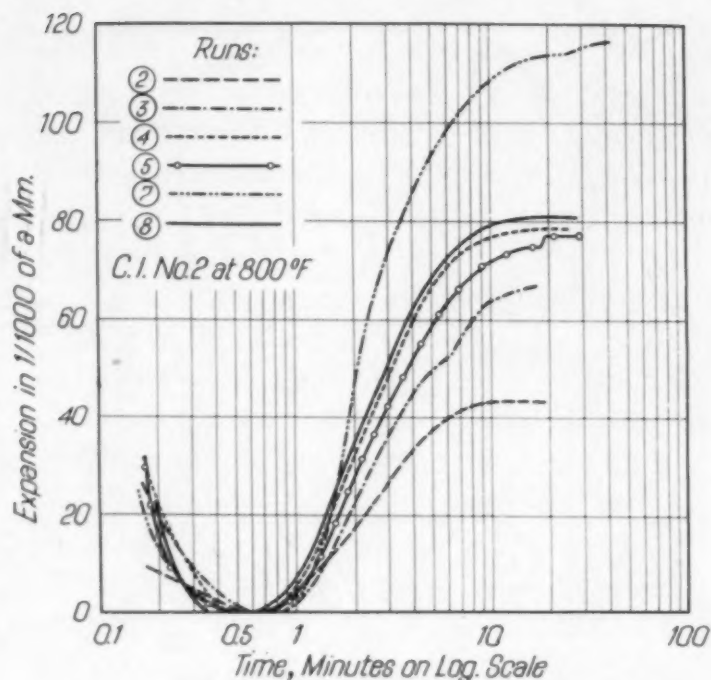


Fig. 10—Expansion-Time Curve of Molybdenum Iron Quenched Into Constant Temperature Bath at 800 Degrees Fahr.

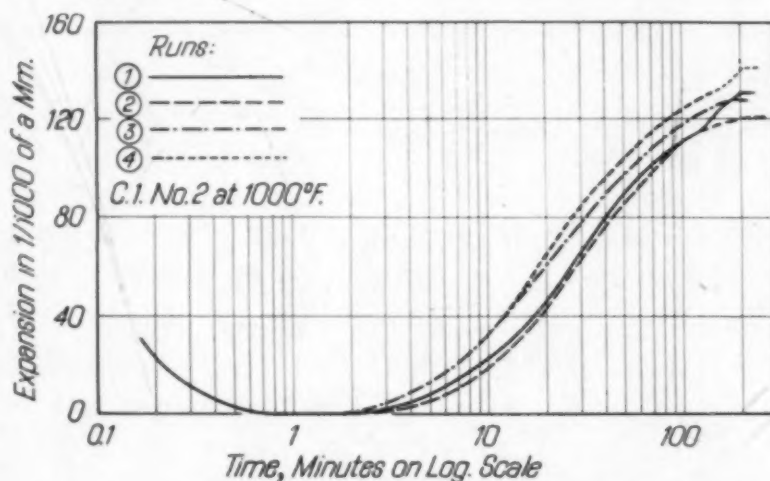


Fig. 11—Expansion-Time Curve of Molybdenum Iron Quenched Into Constant Temperature Bath at 1000 Degrees Fahr.

doubt that they have considerable agreement, and have definite forms for a given set of operating conditions.

The amount of expansion apparently is independent of reaction temperature, because various amounts of expansion were found at any given transformation temperature. An explanation for this may be found in the work of Boyles (4), where he shows that equilibrium diagrams for cast irons of a composition similar to these have an area between 1450 and 1550 degrees Fahr. (790 and 845

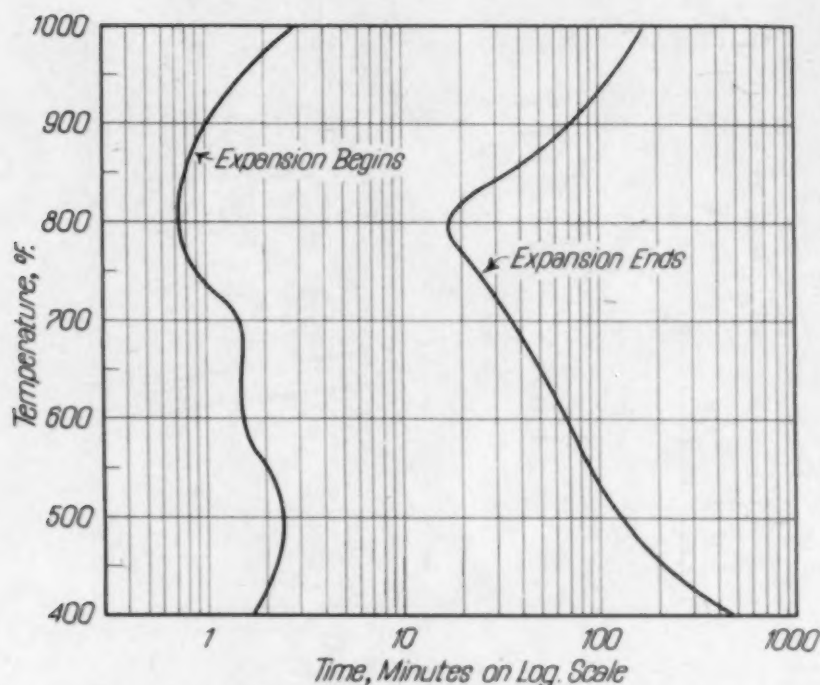


Fig. 12—Time-Temperature Curve for Austenite Reaction in Base Iron from 400 to 1000 Degrees Fahr.

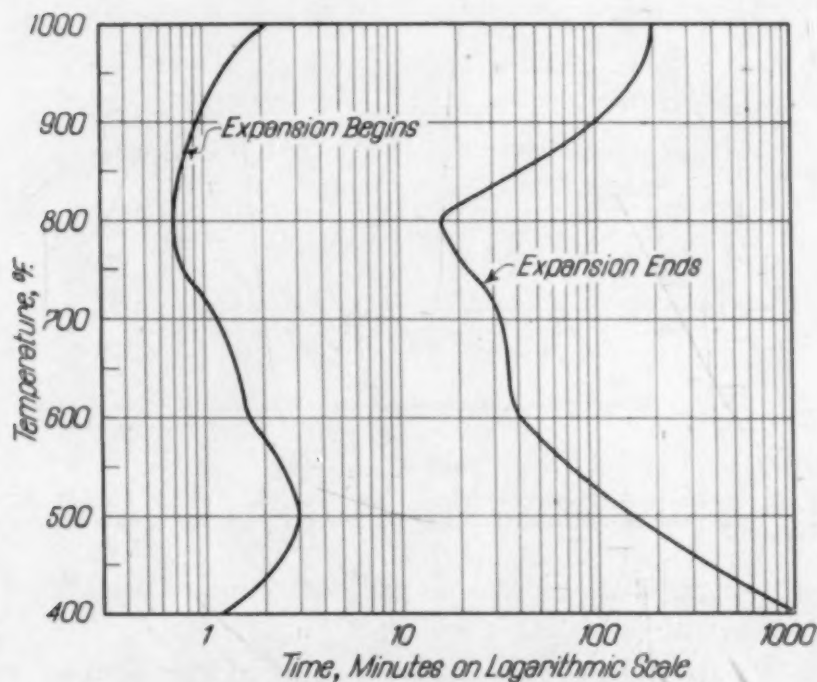


Fig. 13—Time-Temperature Curve for Austenite Reaction in Molybdenum Iron from 400 to 1000 Degrees Fahr.

degrees Cent.) in which ferrite, graphite, and austenite are in equilibrium. Since our irons were heated to 1500 to 1525 degrees Fahr. (815 to 830 degrees Cent.) it is possible that they were not com-

pletely austenitic when quenched. This can be shown later in some photomicrographs of the irons. A small variation in preheating temperature would greatly influence the ferrite content of the iron. Since the amount of austenite present at quenching determines the expansion at constant temperature, any variation in preheating temperature would greatly affect expansion.

The general shape of the curves does seem to have a definite variation with temperature, and is fairly constant for a given temperature and iron. This indicates that the mode of transformation of the austenite is always the same for a given temperature, and that this mode gradually changes with decreasing temperature. This condition is in agreement with austenitic decomposition in steels.

It will be noted that the curves for reactions at the lower temperatures show periods of no expansion, after some primary expansion has taken place. This might be considered as the two-stage reaction alluded to by Davenport and Bain (7), and clearly indicated by Parke and Herzig (15). Also, this type of reaction is not the same in the two irons. The molybdenum iron shows this type of reaction up to 500 degrees Fahr. (260 degrees Cent.), whereas the base iron shows it up to 600 degrees Fahr. (315 degrees Cent.).

The only plausible explanation for irregularities in the curves is that the specimens were not homogeneous. Because of the presence of graphite in the irons there could be large variations in carbon content in the matrix. The segregating tendencies of phosphorus as phosphide eutectic, and of manganese as sulphide inclusions undoubtedly furthered concentration gradients. Variations in the austenite grain size can also cause differences in reaction time (14).

On the summarizing curves, Figs. 12 and 13, the minimum time for the beginning of expansion is 800 degrees Fahr. (425 degrees Cent.) in both cases. This minimum for most steels occurs between 900 and 1000 degrees Fahr. (480 and 540 degrees Cent.). This disagreement is probably because precipitation of cementite at higher temperatures is extensive enough to neutralize expansion from ferrite formation. The curves are quite similar, the most significant difference being at 400 degrees Fahr. (205 degrees Cent.).

Finally, the tremendously increased time required, both for beginning and ending the reactions as compared to most steels, is quite significant. Similar observations have been made by Murphy, Wood and Girardi (14).

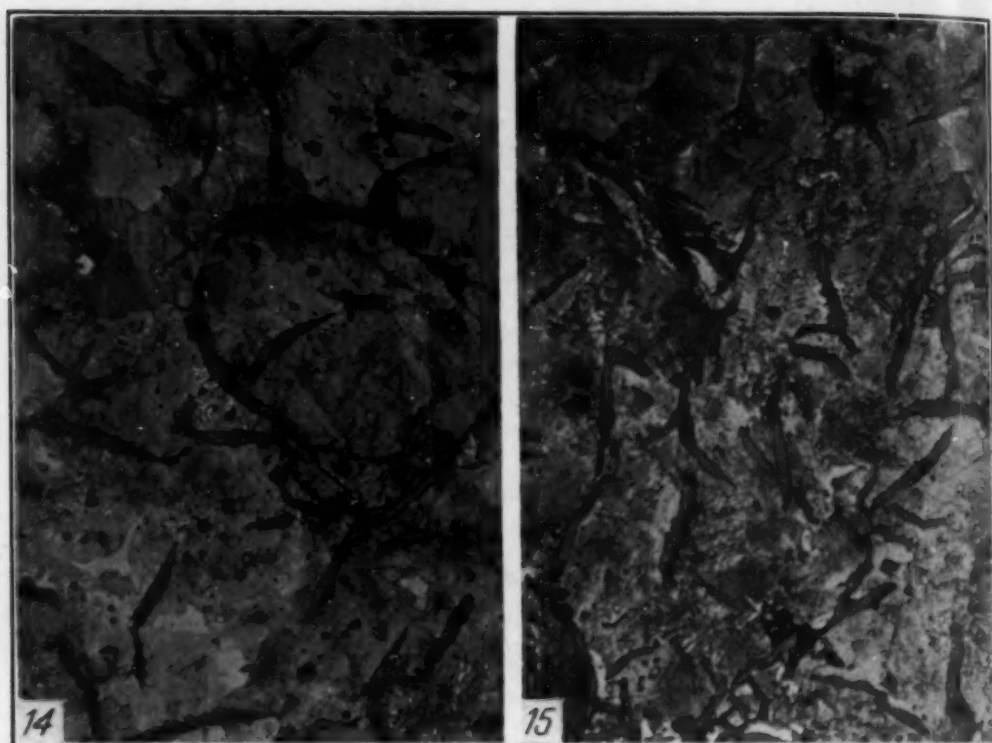


Fig. 14—Base Iron as Cast, 2.9 Per Cent Carbon, 2.1 Per Cent Silicon, 250 Diameters, Nital Etch.

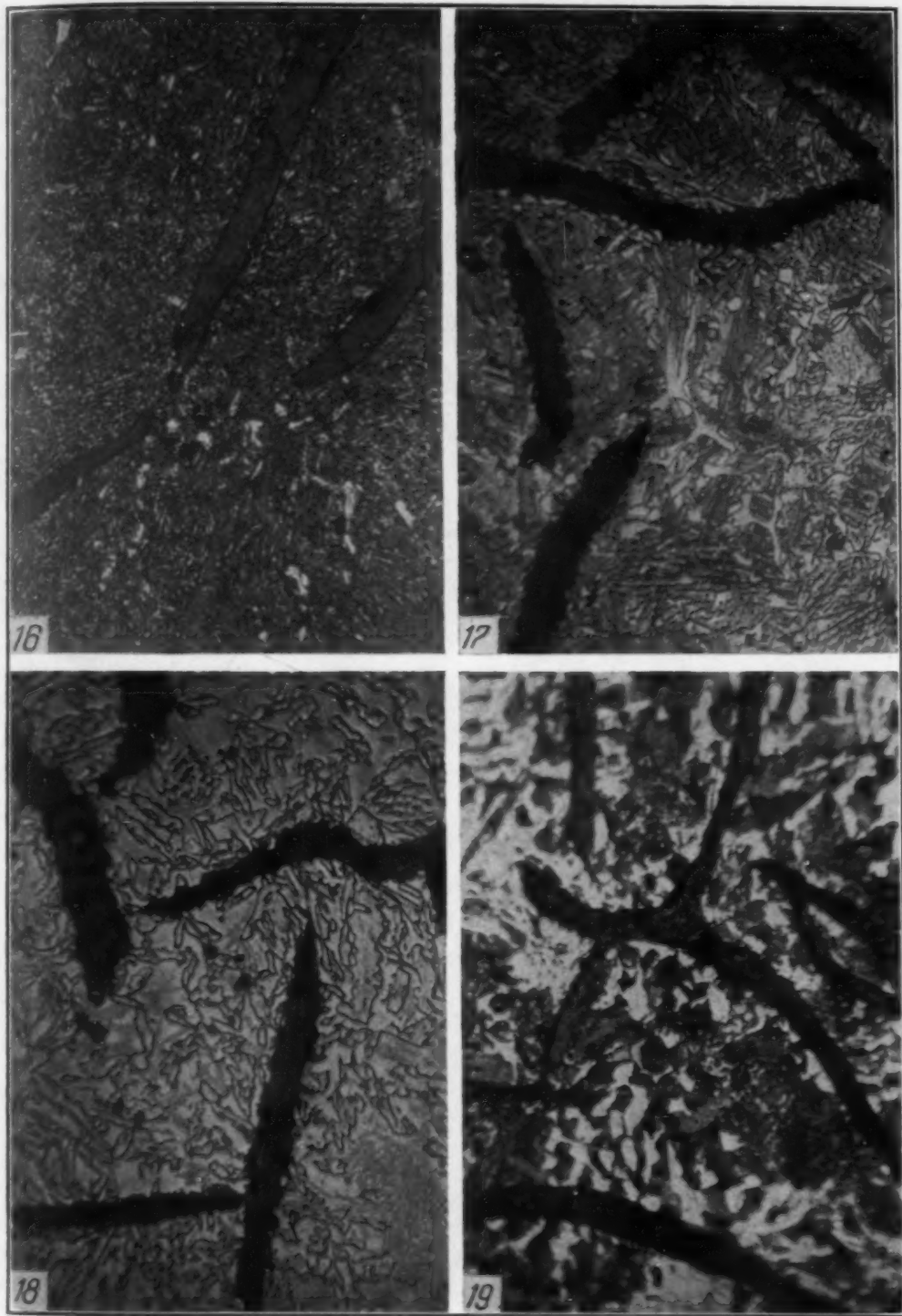
Fig. 15—Molybdenum Iron as Cast, Base Iron Plus 0.50 Per Cent Molybdenum, 250 Diameters, Nital Etch.

PHOTOMICROGRAPHS

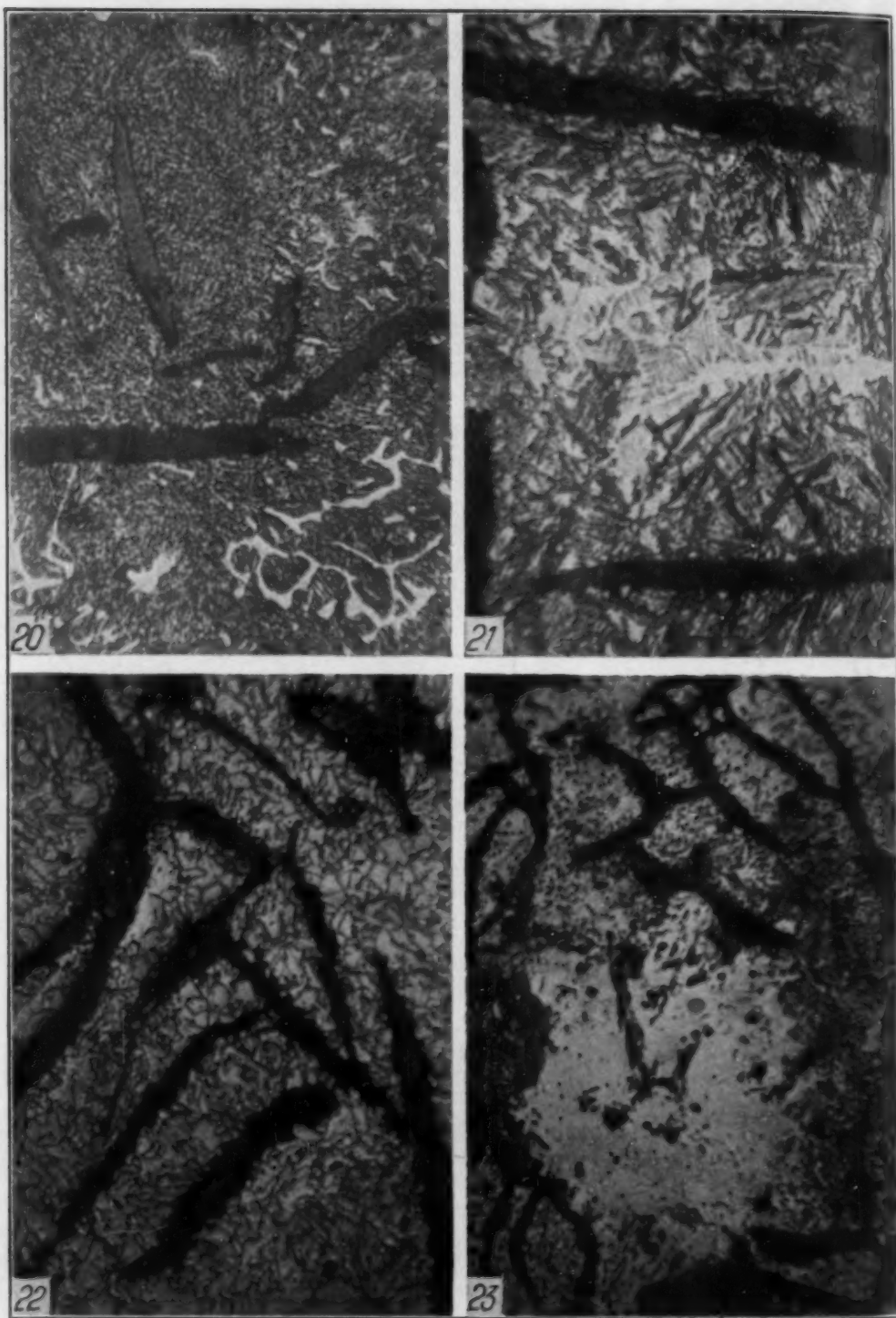
The photomicrographs (Figs. 14 to 23) show the microstructures of the specimens after all expansion was completed. All were etched with 2 per cent nital. The structures in Figs. 14 and 15 are of the original bars as cast, and as they were when heat treatment began. It will be noticed that the matrix of each is almost completely pearlitic. The decreased size of the graphite flakes in the No. 2 iron is generally found in molybdenum irons.

DISCUSSION OF PHOTOMICROGRAPHS

In general, the transformation products from decomposed austenite in these cast irons vary in a manner similar to those in steels. The higher temperatures yielded a coarse, divorced product of ferrite and cementite which might be considered pearlitic. The troostitic or feathery products which typify upper bainite have been suppressed, however. Instead, the products of the intermediate temperatures (700 to 800 degrees Fahr.) are spheroidized. At 600 degrees Fahr.,



Photomicrographs of Base Iron. Fig. 16—After Reaction at 400 Degrees Fahr. Fig. 17—After Reaction at 600 Degrees Fahr. Fig. 18—After Reaction at 800 Degrees Fahr. Fig. 19—After Reaction at 1000 Degrees Fahr. All at a Magnification of $\times 1000$. Nital Etch.



Photomicrographs of Molybdenum Iron. Fig. 20—After Reaction at 400 Degrees Fahr. Fig. 21—After Reaction at 600 Degrees Fahr. Fig. 22—After Reaction at 800 Degrees Fahr. Fig. 23—After Reaction at 1000 Degrees Fahr. All at a Magnification of $\times 1000$. Nital Etch.

and below, the acicular appearance of lower bainite can be noted. It may be seen that this lower bainite becomes progressively finer in texture as reaction temperatures decrease, and approaches the structure of martensite as a limit.

The white areas appearing in the photomicrographs have been interpreted as ferrite. This ferrite was in equilibrium with the austenite and graphite at the 1500 degrees Fahr. preheating stage, as has been shown by Boyles (4). In this case it would have no effect upon the austenite actually present, and would remain as it was throughout the salt bath treatment. This view may be supported by the appearance of remnants of ferrite grains in the photomicrographs, showing that the ferrite had not changed since before it was quenched, when it was being dissolved in austenite. The larger amounts of ferrite in the high temperature products can be expected because the cooling rate during quenching was not as great as in the other cases. This gave more time for ferrite formation from austenite in the ferrite equilibrium range.

CONCLUSIONS

In the final analysis it would appear that 0.5 per cent molybdenum has no great effect upon the reaction rates of this cast iron except at the lowest temperature studied, 400 degrees Fahr. The reaction products, especially at high temperatures, have a finer texture in the molybdenum iron. The general natures of the reaction products in both irons for a given temperature are quite similar. Modes of reaction, as interpreted from the dilation curves, are similar, except that the two-stage reaction at lower temperatures is suppressed somewhat by molybdenum.

ACKNOWLEDGMENT

The authors are grateful for the cooperation and interest shown by the Climax Molybdenum Company in furnishing the specimen material which made this work possible.

Bibliography

1. F. H. Allison, Jr., "Genesis of Pearlite and Martensite," *METAL PROGRESS*, Vol. 28, 1931, p. 60.
2. J. B. Austin, "Dependence of Rate of Transformation of Austenite on Temperature," *Transactions, American Institute of Mining and Metallurgical Engineers, Iron and Steel Division*, Vol. 116, 1935, p. 309.

3. E. C. Bain, "The Process and Result of Austenite Transformation at Constant Temperature," *Metals and Alloys*, Vol. 8, 1937, p. 22.
4. Alfred Boyles, "The Pearlite Interval in Gray Cast Iron," American Foundrymen's Association Preprint No. 40-16, 1940.
5. Sir H. Carpenter and J. M. Robertson, "Metals," Oxford University Press, 1939.
6. W. R. Chapin, "Properties of Some Steels in the Hardening Range," *TRANSACTIONS*, American Society for Steel Treating, Vol. 2, 1922, p. 507.
7. E. S. Davenport and E. C. Bain, "Transformation of Austenite at Constant Subcritical Temperatures," *Transactions*, American Institute of Mining and Metallurgical Engineers, Iron and Steel Division, Vol. 90, 1930, p. 117.
8. B. H. Delong and F. R. Palmer, "What Happens When High Speed Steel is Quenched," *TRANSACTIONS*, American Society for Steel Treating, Vol. 13, 1928, p. 420.
9. H. J. French and O. Z. Klopsch, "Quenching Diagrams for Carbon Steels in Relation to some Quenching Media for Heat Treatment," *TRANSACTIONS*, American Society for Steel Treating, Vol. 1, 1919, p. 469.
10. A. B. Greninger and A. R. Troiano, "Kinetics of the Austenite to Martensite Transformation," *TRANSACTIONS*, American Society for Metals, Vol. 28, No. 3, 1940.
11. H. Hanemann and H. J. Wiester, "Martensite Crystallization in the High Carbon Steels," *Archiv für das Eisenhüttenwesen*, Vol. 5, January, 1932, p. 377.
12. Lewis Dartrey, "The Transformation of Austenite into Martensite in a 0.8 per cent Carbon Steel," *Journal*, Iron and Steel Institute, Vol. 1, 1929, p. 427.
13. R. F. Mehl, "The Physics of Hardenability," Hardenability of Alloy Steels, Published by American Society for Metals, 1939, p. 1.
14. D. W. Murphy, W. P. Wood and D. Girardi, "The Rate of Austenite Transformation in Cast Iron," *TRANSACTIONS*, American Society for Metals, Vol. 26, 1938, p. 173.
15. R. M. Parke and A. J. Herzig, "Hardenability of Molybdenum S. A. E. Steels," *Metals and Alloys*, Vol. 1, 1940, p. 6.
16. A. M. Portevin and M. Garvin, "Investigation of the Influence of the Rate of Cooling on Hardening of Carbon Steels," *Journal*, Iron and Steel Institute, Vol. 1, 1919, p. 469.
17. G. B. Upton, "The Habits and Laws of Decomposition of Super-Cooled Solutions with Special Regard to Austenite," *TRANSACTIONS*, American Society for Metals, Vol. 22, 1934, p. 690.
18. F. Wever and W. Jellinghaus, "On the Influence of Chromium on the Transformation of Carbon Steels," *Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung*, Vol. 14, 1932, p. 105.
19. F. Wever and H. Lange, "The Transformation Kinetics of Austenite," *Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung*, Vol. 14, 1932, p. 71.
20. S. Steinberg and V. Sjusin, "Transformation of Austenite in High-Chromium Steel," *Technical Physics of the U. S. S. R.*, Vol. 1, 1935, p. 376.
21. Handbook of Chemistry and Physics, 21st ed., p. 1255.
22. Metals Handbook, 1939 ed., American Society for Metals, p. 425.

DISCUSSION

Written Discussion: By J. E. Schoen, professor and head, department of mechanical engineering, Marquette University, Milwaukee.

I have read the paper by Messrs. Oakley and Oesterle and was much interested in the results of the investigation. They have contributed considerable

information to the literature on this subject, and it has always been my feeling that the more we can learn about the transformation of austenite, the better able we will be to draw conclusions. Austenite whether free or an admixture with other constituents such as are found in cast iron should transform into the various transformation products modified only as it is saturated or unsaturated with other constituents. The dilatation curves definitely indicate the effects of some constituents on this transformation. It would be interesting to see this study proceed with greater control of the substances by which the mechanism of transformation is varied. It is interesting to note that the time required for the transformation is constant regardless of the amount of dilatation obtained. I rather like their ingenuous method of devising their own type of test specimen, and the simplicity which it lends to the study.

Written Discussion: By C. A. Nagler, instructor, and R. L. Dowdell, professor of metallurgy, University of Minnesota, Minneapolis, Minn.

The authors are to be congratulated on the design of their dilatometer and the apparent serviceability of it. The writers have given consideration to a device of the type of the Oesterle dilatometer but never got to the point of making one for experimental use.

It is rather unfortunate, however, that the authors did not hold these cast irons at higher temperatures, considerably above 1500 to 1525 degrees Fahr.,

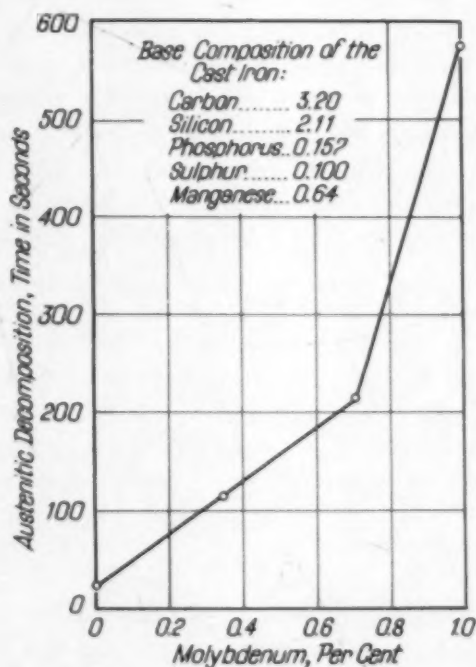


Fig. 1—Effect of Molybdenum on the Time for Transformation of Austenite at 1200 Degrees Fahr.

so that more of the carbide phase would be in solid solution and thereby obtain a more stable austenite having a slower rate of decomposition.

We are wondering whether or not the authors used the same samples for their transformation data at each temperature as given in Figs. 4 to 11 inclu-

sive, and also how they determined the beginning and ending of the austenite transformation. We realize this to be a difficult problem and variations will occur with different methods such as dilatometric, microscopic or magnetic.

The authors conclude "In the final analysis it would appear that 0.5 per cent molybdenum has no great effect upon the reaction rates of their cast iron except at the lowest temperature studied, 400 degrees Fahr." The writers assume the authors mean that a molybdenum addition of 0.5 per cent to a base iron of 2.9 per cent carbon with 2.1 per cent silicon does not affect the rate of austenitic decomposition. Fig. 1 submitted in this discussion shows that practically the same base iron made synthetically by the same company, Climax Molybdenum Corp., in an induction furnace, had a faster rate of reaction than the base iron with additions of 0.35, 0.71 and 1.00 per cent molybdenum at 1200 degrees Fahr. Of all the alloy elements used in cast iron to delay the decomposition of austenite, molybdenum is without a doubt the most effective in quantities up to 1 per cent.

It is unfortunate that full details of the magnetic method used by the writers cannot be injected here into the discussion but these details will probably be published shortly by A.S.M. under the authorship of W. P. Wood and C. A. Nagler.

Written Discussion: By George A. Timmons, metallurgist, Climax Molybdenum Company of Michigan, Detroit.

The authors' conclusion concerning the effect of molybdenum on the rate of transformation of austenite in cast iron conflicts with accepted knowledge that alloying elements (except cobalt) increase the austenite stability of iron carbon alloys at temperatures above 900 degrees Fahr. One is not surprised to find little difference in the rates of reaction in the two irons at temperatures below 900 degrees Fahr. since the study of austenite decomposition in steels has shown usual alloy additions to have relatively little effect at lower temperatures.

Murphy, Wood and D'Amico¹ have shown the region between 1200 and 1000 degrees Fahr. to be the significant part of the S-curve for cast iron. It is unfortunate that the authors did not investigate this temperature range.

Although the authors have not drawn conclusions regarding the effect of molybdenum on the hardenability of cast iron, from their data at 1000 degrees Fahr. one might be led to infer that molybdenum does not contribute to the hardenability of cast iron.

Excellent correlation between S-curves and hardenability tests have been obtained by others.^(2, 3) On the basis of the authors' data, the two irons studied in this paper should, therefore, exhibit similar hardenability curves when quenched from the same austenitizing temperature. Hardenability tests fail to show this similarity. Jominy⁴ hardenability specimens were prepared

¹D. W. Murphy, W. P. Wood, and C. D. D'Amico, "Austenite Transformation in Gray Iron," *Transactions, American Foundrymen's Association*, 1938, p. 563-586.

²R. M. Parke and A. J. Herzig, "Hardenability of Molybdenum S.A.E. Steels", *Metals and Alloys*, Vol. 1, 1940, p. 6.

³D. W. Murphy, W. P. Wood and D. Girardi, "The Rate of Austenite Transformation in Cast Iron," *TRANSACTIONS, American Society for Metals*, Vol. 26, 1938, p. 173.

⁴W. E. Jominy, "Hardenability Tests", *Hardenability of Alloy Steels*, published by American Society for Metals, 1939, p. 66.

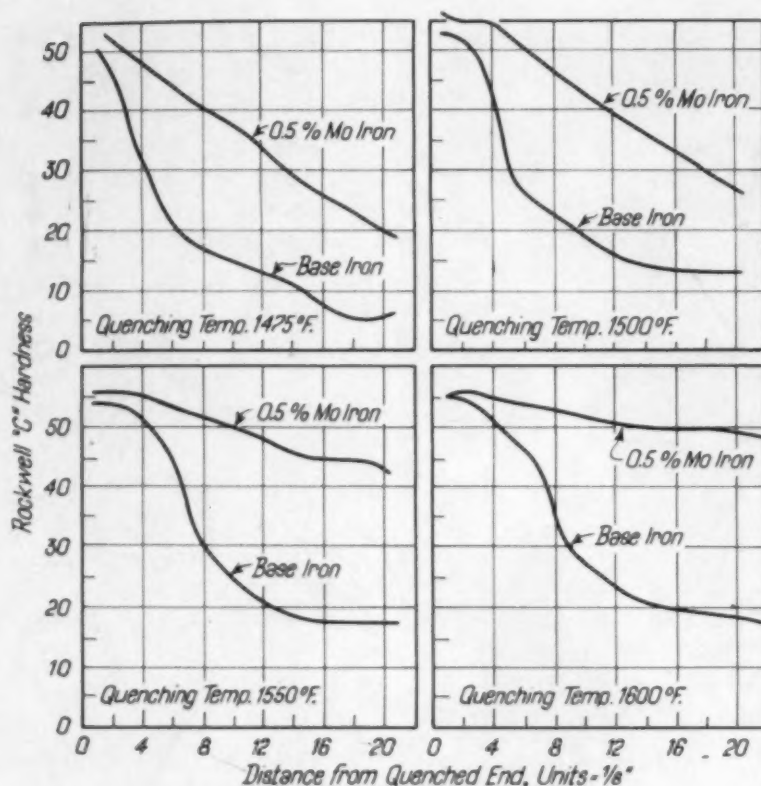


Fig. 2—Hardenability Determinations Using the Jominy Specimen From Samples of the Same Heats of Irons Used by the Authors and Quenched from Four Different Temperatures.

from samples of the same heats of irons used by the authors and were quenched from four different temperatures: 1475, 1500, 1550 and 1600 degrees Fahr. Fig. 2 shows the results of these hardenability determinations. Note that at each temperature the 0.5 per cent molybdenum iron showed a superior hardenability rating.

The authors' S-curve for the base iron seriously varies with those obtained by Murphy, Wood and D'Amico. The latter showed S-curves for two similar irons and revealed that the time required for initiation of austenite decomposition at 1000 degrees Fahr. was 7 to 10 seconds compared with some 160 seconds as indicated on the authors' curve.

A metallographic study of the rates of reaction of the two irons at 1000 degrees Fahr. also failed to support the authors' conclusion. Fig. 3 shows comparative microstructures of the two irons produced by quenching specimens, 0.063 inches thick, from 1525 degrees Fahr. into a lead bath at 1000 degrees Fahr. where they were held 300 seconds before quenching into water. These photomicrographs indicate that 90 per cent of the austenite of the plain iron had decomposed in this time compared with only 7 per cent of the austenite in the molybdenum iron.

It is believed that the disagreements involved are due primarily to the fact that the specimen in the dilatometer cools from the austenitizing temperature to that of the quenching bath at a rate less than the critical cool-

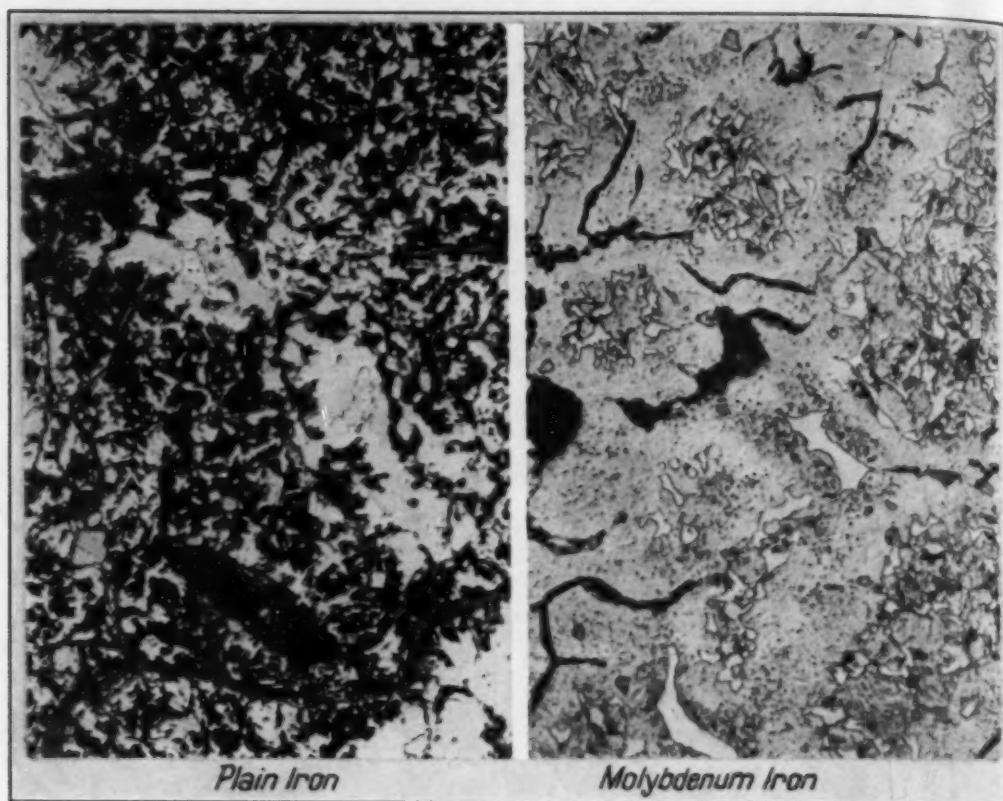


Fig. 3—Microstructures of Plain Iron and 0.5 Per Cent Molybdenum Iron. Quenched from 1525 Degrees Fahr. into Lead at 1000 Degrees Fahr., Held 300 Seconds, Quenched into Water.

ing rate and, therefore, the time-expansion curves obtained do not indicate the actual start of austenite decomposition.

The authors' quenching dilatometer possesses a serious disadvantage for accurate determinations of this sort. The inside tube is sealed off at the bottom by the fused silica button, preventing rapid cooling of this tube by the quenching medium. The rate of quench is, therefore, considerably slower than desirable for determinations where the rate of reaction is apt to be very rapid. The outside of the specimen may be cooled rapidly by its contact with the salt bath used for quenching, but the inside of the specimen, adjacent to the inner quartz tube, will remain at a higher temperature for some time due to the slow rate of cooling of the tube whose heat conductivity is extremely low.

An interpretation of the time-expansion curves must take into account that there are three possible motions which affect the shape of this curve: (a) thermal contraction of the specimen due to its cooling from the austenitizing temperature to that of the quenching medium; (b) expansion caused by the allotropic change; (c) thermal contraction of the inner quartz tube over the 2-inch length enclosed by the specimen.

An arrest in the motion of the dial does not, therefore, necessarily mean that the specimen has reached the bath temperature nor does an apparent expansion indicate that transformation has begun.

At 1000 degrees Fahr. the base iron showed a total expansion of from 15

to 33 units while the molybdenum iron showed a total expansion of from 120 to 140 units, suggesting that in the case of the plain iron some austenite had decomposed before the specimen reached the temperature of the quenching bath. The microstructure of the base iron transformed at 600 degrees Fahr. shows two distinct types of decomposition products which appear to be more the result of differences in temperature of transformation than failure to eliminate ferrite at the austenitizing temperature.

Evidence has been presented in this discussion to show that 0.5 per cent molybdenum materially retards the rate of transformation of cast iron at 1000 degrees Fahr. and increases hardenability.

Oral Discussion

J. J. KANTER:⁵ I would like to ask the authors how the dilatometer distinguishes between dilation which is due to transformation and dilation which may be due to graphitization. I believe the compositions they worked with are liable to graphitize to a certain extent in the temperature ranges at which transformation occurs and thus render the data indeterminate as to the extent and rate of either transformation or graphitization.

Authors' Reply

In reply to general criticism, the authors realize that some of the data presented does not conform with data already published on this material. It is just possible that we have not interpreted our data correctly.

We wish to thank each of the gentlemen presenting discussions for the interest they have taken. We are in entire agreement with the comments made by Professor Schoen.

We also concur to a large extent on the comments made by Messrs. Dowdell and Nagler. We did not heat our specimens preliminary to quenching to temperatures above the range, 1500 to 1525 degrees Fahr., in order to avoid as much as possible complications of the added variable of the higher carbon austenite. We felt that we eliminated one variable by using a new specimen for each expansion curve and by quenching the same from the same temperatures after the dilatometer gave reasonable assurance that the dimension was stationary.

Our curves do show some retarding effect due to the presence of molybdenum, but this was not of great magnitude. We were careful in the paper to refer only to the beginning and ending of expansion, and, therefore, we cannot insist that the data presented can be interpreted in the light of transformation. We feel you must grant, however, that there is nice conformity in the beginnings and endings of expansion.

Mr. Timmons had some very interesting comments. The many variables in cast iron make this study rather complicated. No conclusions, therefore, should be drawn from dilation studies alone, unless one is using the method to obtain a general picture of one particular alloy. Since his comments show that molybdenum produces the best results above 1000 degrees Fahr. and our

⁵Research metallurgist, Crane Co., Chicago.

work started at 1000 degrees Fahr., they might be considered contributory rather than critical. We selected the tubular specimen in order to avoid the insulating effect of fused silica. Our original dilatometer was intended for the slower heating and cooling thermo-analyses and used a solid bar. More than half of our tubular specimen was exposed to the quenching action of the bath, and we felt the point raised by Mr. Timmons to be negligible. We have no way of knowing at this time that our quenching rate to the austenizing temperature was less than critical. We do know of some nice austempering that has been done on molybdenum cast irons in which this same kind of bath was used. The variations we obtained in the actual expansions of the specimens are still an unknown quantity with us, but they have no bearing on this paper.

In regard to Mr. Kanter's question, we did not distinguish between the factors contributing toward dilation. As mentioned before in the closure, all specimens were heated to the same temperature, approximately 1525 degrees Fahr., held until the gage showed no further dilation, and then quenched into a liquid at the austenizing temperature. It is possible that some of our nonuniform expansion curves might have an element of graphitization in their make-up. We recorded only the over-all effect.

QUANTITATIVE EVALUATION OF DISTORTION IN SILICON STEEL AND IN ALUMINUM

BY G. L. CLARK AND W. M. SHAFER

Abstract

In continuation of previous work by Clark and Beckwith and by Clark and Dunn on the X-ray diffraction study of distortion in metals subjected to various types of stress, the attempt is made to establish a considerably more quantitative correlation between measured stresses and observed effects on X-ray patterns. In a carefully designed apparatus, strips of silicon steel made up of a single crystal were bent in such a way that the radius of curvature could be accurately measured by a modified spherometer and the X-ray pattern made in this position. The elongation of Laue spots was measured with the aid of Leonhardt curves and correlated directly with the bending curvature of the specimen. The stress applied below the elastic limit was then calculated with appropriate and tested equations. Under the conditions of these experiments, radial asterism is observed within the elastic limit when single crystals of silicon steel are bent, and beyond the elastic limit a part of the rotation or bending of the planes is elastic. Results are obtained also for single crystals of aluminum and for polycrystalline specimens subjected to bending, rolling and elongation by tension. Equivalent amounts of cold working when expressed in terms of Nadai's "octahedral shear," whether by rolling or elongation by tension within the limits of deformation studied, produce a similar effect as revealed by the X-ray patterns. This definite correlation establishes the general validity of the theoretical concept of octahedral shear. The mechanism of deformation under each type of stress is established from the X-ray patterns.

MUCH work of a qualitative nature has been done on the X-ray study of crystalline distortion. Because of the variables and complexities encountered, not much quantitative work has as yet been done. As this investigation was not concerned primarily with theories and mechanisms of distortion, a discussion of the work bearing on these will not be undertaken here. Such discussions have been ably

A paper presented before the Twenty-second Annual Convention of the Society held in Cleveland, October 21 to 25, 1940. Of the authors, G. L. Clark is professor of chemistry, University of Illinois, Urbana, Ill., and W. M. Shafer is assistant professor of chemistry, Iowa State Teachers College, Cedar Falls, Iowa. Manuscript received July 1, 1940.

conducted by Clark and Beckwith (1a),¹ by Clark and Dunn (1b), by Hoyt (2), and by Gough (3).

Distortion or strain in crystals is manifested in the X-ray diffraction pattern by radial streaks or asterism; by radial widening of lines; by shifting of lines and spots; by peripheral widening of spots; and by changes in intensities of reflections. Any quantitative work must correlate one or more of these pattern manifestations with stress or strain.

Some of the first work done on the lengthening of Laue spots caused by distortion as that of Rinne (4) in 1924. He studied patterns of bent rock salt. Leonhardt, (5) also in 1934, calculated the curves along which the Laue spots should elongate for a given axis of bending and orientation of the crystal. These curves have proven useful in this work and will be referred to later.

Goucher (6) extended large crystal and single crystal tungsten wires and measured the peripheral widening versus per cent reduction in diameter. While the results showed much scattering, it was clear that at least an approximate linear relationship was present from 0-40 per cent reduction in diameter. Regler (7) measured the peripheral widening of spots for large-angle diffraction lines as a function of the amount of deformation and claimed great accuracy for this method.

Several series of diffraction patterns of metals subjected to known strains and stresses have been published. Such a series has been published by Clark (8) for a single crystal of silicon steel subjected to known tensile stresses. Froumer (9) found it possible by means of X-ray diffraction patterns to establish in a material of otherwise known metallurgical history the amount of cold work imparted to it by comparing its diffraction pattern with those of a series of pieces of similar material cold-worked in a similar way.

Wood (10) working with metal of sufficiently small grain size to give smooth Debye-Sherrer rings found that line breadth increased progressively with increasing reduction by drawing until it reached a definite maximum which was characteristic of the metal. Brindley and Ridley (11) have correlated lattice distortion in fine-grained copper and nickel with observed changes in the widths and intensities of X-ray reflections.

Clark and Beckwith (12) devised a modulus based upon the ratio of the longest dimension of a Laue spot divided by its shortest

¹The figures appearing in parentheses pertain to the references appended to this paper.

dimension which served as a measure of slight residual distortion in silicon steel. Recently, Jacquesson (13) measured quantitatively the effect of known torsions upon cylindrical crystal specimens of aluminum. He measured the angular displacement of the crystallites as shown by the X-ray diffraction pattern.

Bollenrath and Osswald (14), working with unalloyed, low carbon steel, measured the internal stress from the change in lattice parameters, as revealed by the X-ray diffraction pattern.

CORRELATION OF BENDING AND STRESS IN STRIPS OF LARGE CRYSTALS OF SILICON STEEL WITH ELONGATION OF THE LAUE SPOTS

Preparation of specimens—The specimens used were furnished by Doctor C. G. Dunn, who prepared them by a strain and annealing process. They were 3 per cent silicon steel and the crystals averaged approximately 0.7 square centimeters in area. They were as thick as the specimen. The scale was removed by pickling in warm, dilute sulphuric acid. The specimen used was machined such that it was 0.018 centimeters thick, 0.570 centimeters wide, and 7 centimeters long. These dimensions permitted a single crystal to extend across the full width.

Apparatus for bending and measurement of bending and stress—The specimen was bent by the device shown in Fig. 1B. This apparatus was designed and made to include a collimating tube which fitted into the usual collimating tube support. The specimen was slipped over the two knife edges on opposite sides of the collimating tube and under the two end knife edges. Bending was produced by supporting the apparatus in an inverted position and placing weights in a pan suspended by means of metal hooks from the center of each side of the piece containing the two middle knife edges. This piece is free to move up and down. Shafts running through the center of the milled heads fit loosely into the knife edge piece and prevent it from rotating. The milled heads support the knife edge pieces.

Weight applied to the pan causes equal known forces to be exerted by the knife edges upon the specimen. The specimen was locked in the bent position produced by the known forces by screwing down the milled heads until they barely made contact with the knife edge piece. With the specimen locked in its bent position, the weights

and pan were removed and the apparatus could be turned to any position without disturbing the specimen.

With forces applied in this manner, the bending was circular because the bending moment was constant throughout the specimen for a given weight. The bending was strictly circular only within the elastic limit of the specimen.

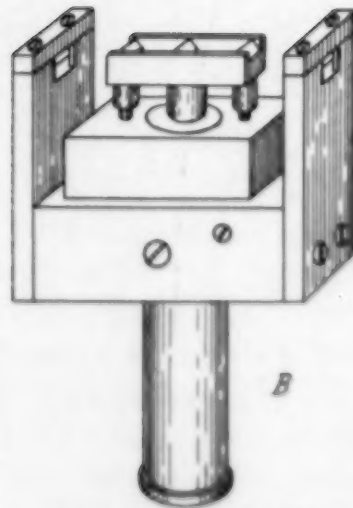
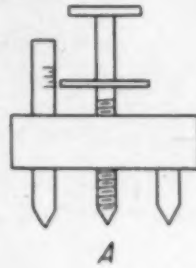


Fig. 1A — Modified Saha.

specimen to film distance of 4 centimeters. The length of the collimating tube was 3.5 inches and the pinhole diameter was 0.25 inches. Since the elongation of a Laue spot depends upon the divergence of the X-ray beam and the area of the crystal struck by the beam, as well as the radius of curvature of the reflecting planes, all conditions except the radius of curvature had to be kept constant.

Molybdenum polychromatic radiation was used with the tube

Table I
Silicon Steel

Experiment	Figure	Curvature in Re- ciprocal cms.	$\Delta \delta$ Rotation of Laue Spot in Degrees	Maximum Stress lbs. per in. ²	Comments
D ₁	2	0	0.45	0	zero reading
D ₄	..	0.055	0.40	$17.9 \times 10^{+3}$	
D ₆	..	0.104	1.05	$27.5 \times 10^{+3}$	
D ₁₁	3	0.1525	1.30	$40.75 \times 10^{+3}$	
D ₁₄	..	0.191	1.50	$49.9 \times 10^{+3}$	
D ₂₀	4	0.2575	2.05	$64.5 \times 10^{+3}$	first appearance of residual bending
D ₂₁	5	0.0185	1.00	0	
D ₂₃	6	0.433	5.25	not measured	residual bending
D ₂₆	7	0.159	3.62	0	
D ₂₄	8	0.498	5.80	not measured	
D ₂₅	9	0.206	4.50	0	
D ₂₇	10	0.562	7.50	not measured	
D ₂₈	11	0.263	4.65	0	residual bending

operating at 30 kilovolts and 15 milliamperes. The exposure time was 40 minutes. The patterns are shown in Figs. 2-11, inclusive.

The conditions of bending and stress under which the patterns were registered are given in Table I. In most cases, after a pattern was registered with the specimen in a forced position, a second pattern was registered after external forces had been removed. Any residual bending of the specimen was also measured by means of the modified spherometer.

Orientation of crystal. The orientation of the crystal used was

and pan were removed and the apparatus could be turned to any position without disturbing the specimen.

With forces applied in this manner, the bending was circular because the bending moment was constant throughout the specimen for a given weight. The bending was strictly circular only within the elastic limit of the specimen.

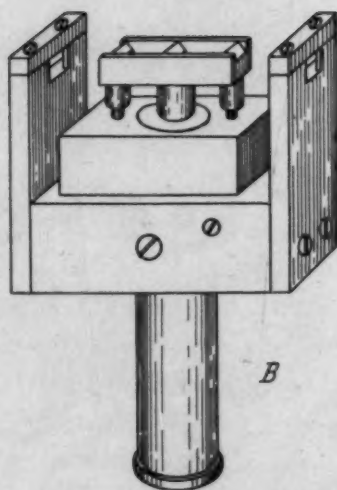
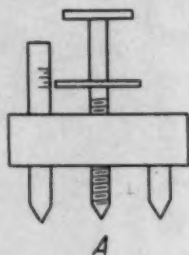


Fig. 1A — Modified Spherometer Used to Measure Curvature of Bent Metal Specimen.
Fig. 1B—Apparatus for Producing Circular Bending of Metal Strip, Combining Collimating Tube for Defining X-ray Beam.

The amount of bending was measured by the use of the device constructed as shown in Fig. 1A. This device, which is a modified spherometer, permitted accurate measurements of the amount of bending to be made. In each case, the radius of curvature was calculated.

Registration of Laue patterns—Laue patterns were taken with a

specimen to film distance of 4 centimeters. The length of the collimating tube was 3.5 inches and the pinhole diameter was 0.25 inches. Since the elongation of a Laue spot depends upon the divergence of the X-ray beam and the area of the crystal struck by the beam, as well as the radius of curvature of the reflecting planes, all conditions except the radius of curvature had to be kept constant.

Molybdenum polychromatic radiation was used with the tube

Table I
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D ₆	..	0.104	1.05	27.5x10 ⁺³	
D ₁₁	3	0.1525	1.30	40.75x10 ⁺³	
D ₁₄	..	0.191	1.50	49.9x10 ⁺³	
D ₂₀	4	0.2575	2.05	64.5x10 ⁺³	
D ₂₁	5	0.0185	1.00	0	first appearance of residual bending
D ₂₂	6	0.433	5.25	not measured	
D ₂₃	7	0.159	3.62	0	residual bending
D ₂₄	8	0.498	5.80	not measured	
D ₂₅	9	0.206	4.50	0	
D ₂₇	10	0.562	7.50	not measured	
D ₂₈	11	0.263	4.65	0	residual bending

operating at 30 kilovolts and 15 milliamperes. The exposure time was 40 minutes. The patterns are shown in Figs. 2-11, inclusive.

The conditions of bending and stress under which the patterns were registered are given in Table I. In most cases, after a pattern was registered with the specimen in a forced position, a second pattern was registered after external forces had been removed. Any residual bending of the specimen was also measured by means of the modified spherometer.

Orientation of crystal—The orientation of the crystal used was determined by comparison of the Laue pattern obtained with the standard patterns published by Magima and Togino (15). The (100) face was inclined approximately 25 degrees from the vertical plane containing the X-ray beam and approximately 10 degrees from the horizontal plane containing the X-ray beam. The indices of the Laue spots were also obtained from these patterns.

Measurement of elongation of Laue spot—The elongation of the Laue spots were measured in degrees by the use of some equations and curves worked out by Leonhardt (5). It is assumed that the elongation of the Laue spot is due to a rotation or bending of the

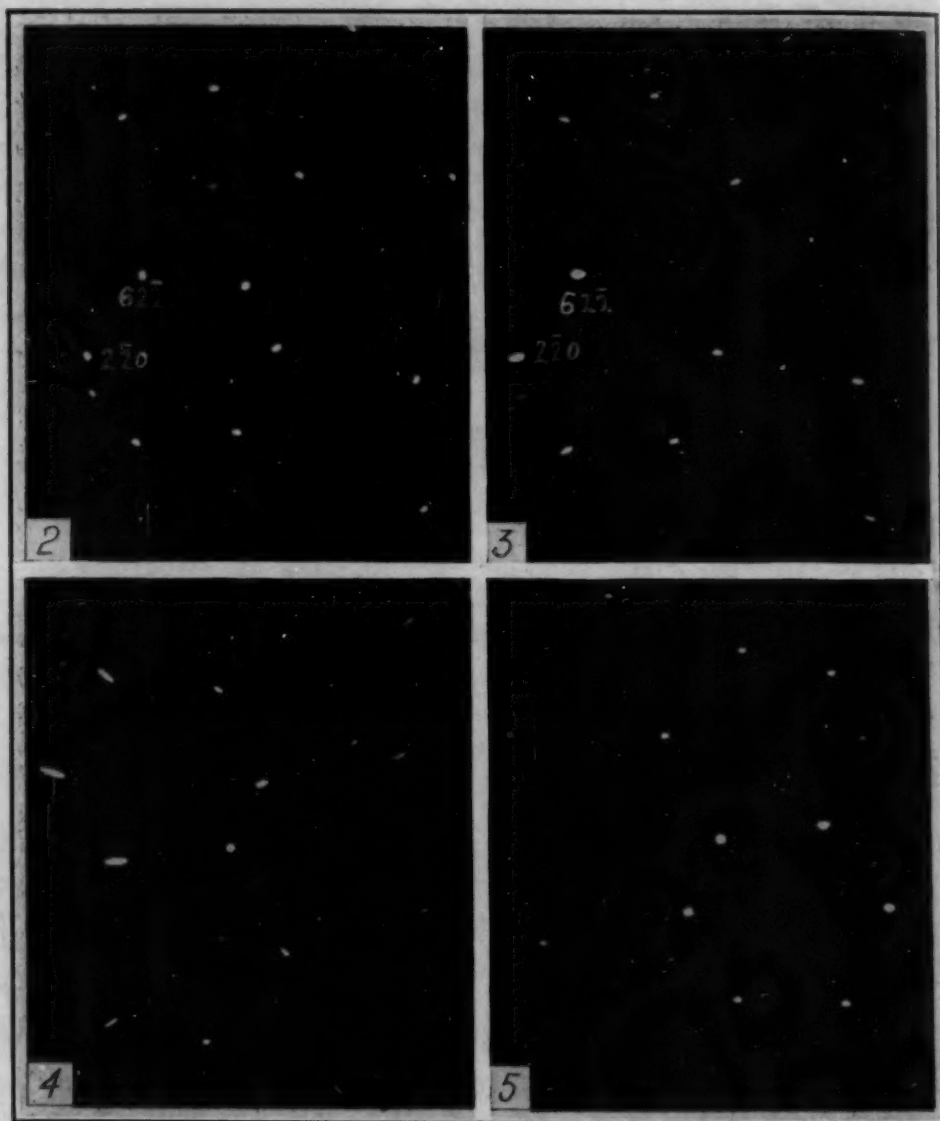


Fig. 2—Laue Pattern of Original Unbent Silicon Steel Crystal.
 Fig. 3—Same Crystal with Curvature of 0.1525 Cm^{-1} (Table I).
 Fig. 4—Same Crystal with Curvature of 0.2575 Cm^{-1} .
 Fig. 5—Same Crystal with Residual Curvature of 0.0185 Cm^{-1} After Removal of Bending Stress.

crystal planes about an axis. The equation used is

$$(1) \quad \sin \delta = \frac{\sin \frac{\theta}{2}}{\sin \sigma}$$

where

$\frac{\theta}{2}$ = Bragg angle

σ = angle between perpendicular to reflecting plane and axis about which the crystal is bent

δ = amount crystal is rotated from the zero position

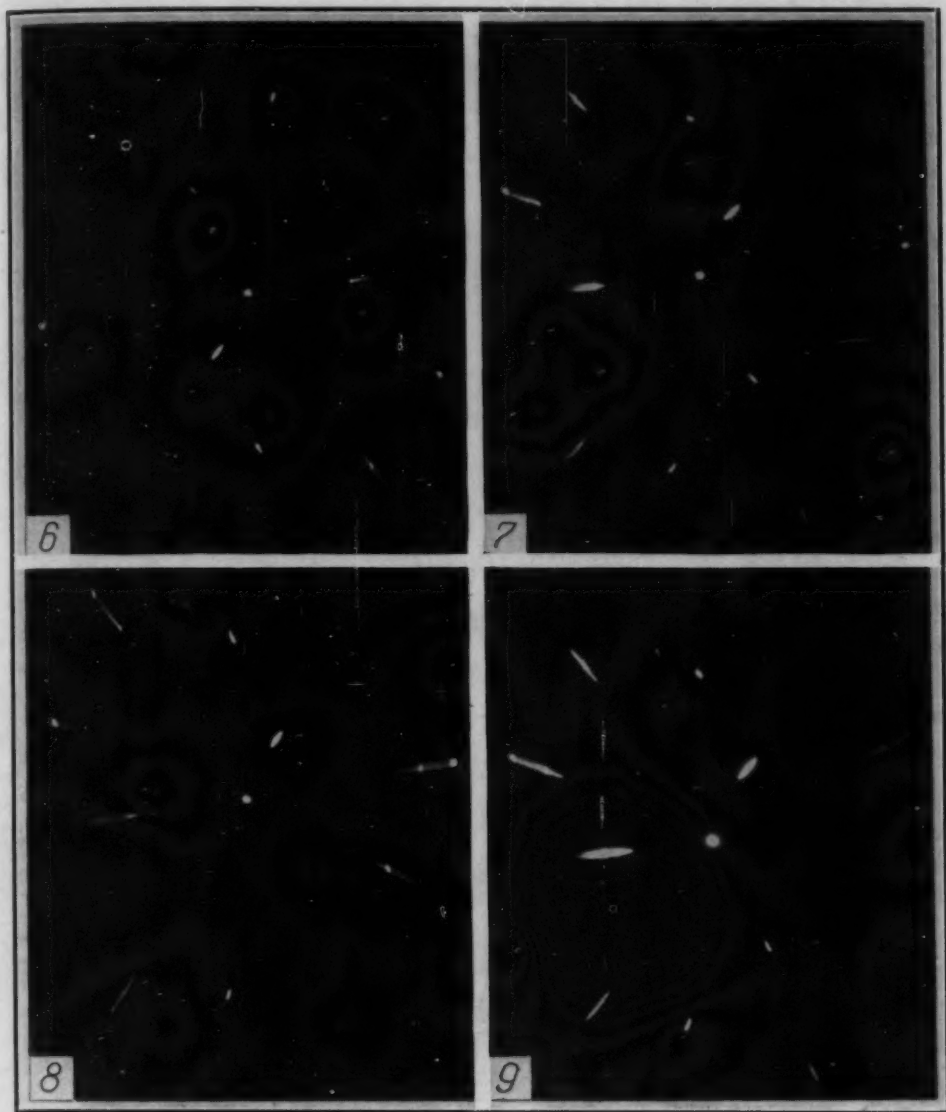


Fig. 6—Same Crystal with Curvature of 0.433 Cm^{-1} .

Fig. 7—Same Crystal with Residual Curvature of 0.159 Cm^{-1} After Removal of Stress in Fig. 6.

Fig. 8—Same Crystal with Curvature of 0.498 Cm^{-1} .

Fig. 9—Same Crystal with Residual Curvature of 0.206 Cm^{-1} After Removal of Stress in Fig. 8.

The elongation of the Laue spot is due to $\Delta\delta$, where $\Delta\delta = \delta_2 - \delta_1$. $\Delta\delta$ was measured and used as a measure of the elongation of the Laue spot. δ_1 and δ_2 were calculated from equation (1).

$$\frac{\theta_1}{2} = \text{Bragg angle for end of elongated Laue spot closest to central spot}$$

$$\frac{\theta_2}{2} = \text{Bragg angle for end of Laue spot farthest from central spot}$$



Fig. 10—Same Crystal with Curvature of 0.562 Cm^{-1} .

Fig. 11—Same Crystal with Residual Curvature of 0.263 Cm^{-1} After Removal of Stress in Fig. 10.

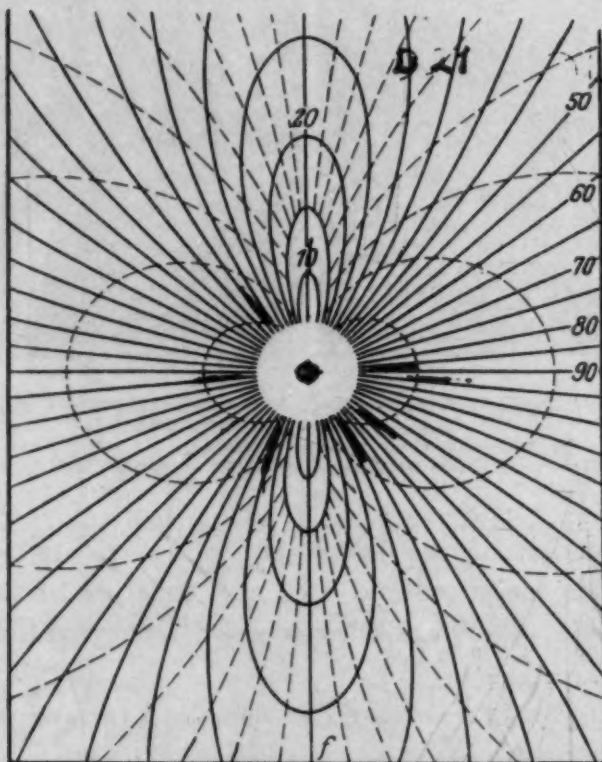


Fig. 12—Leonhardt Curves for Interpretation of Elongation of Laue Spots Under Stress.

σ was found by superimposing the pattern upon Leonhardt's curves (Fig. 12). $\Delta\delta$ was calculated for two Laue spots in each pattern and the average of the two used. These were the reflections from the $62\bar{2}$ and $2\bar{2}0$ planes. Care must be taken in choosing a spot

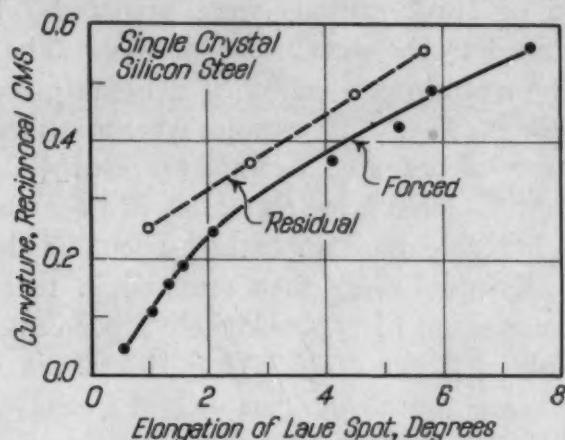


Fig. 13—Elongation of Laue Spots, in Degrees, Plotted Against Curvature of Single Crystal of Silicon Steel.

so that it will not be limited by the short wavelength limit or wavelength of insufficient intensity. The calculated values for $\Delta\delta$ are shown in Table I. $\Delta\delta$ is plotted against the radius of curvature in Fig. 13.

Calculation of stress—The maximum stress on the cross section of the beam was calculated from the equation

$$(2) \quad S = \frac{CM}{I}$$

where:

C = distance from neutral axis (center of specimen in this case) to edge

M = bending moment

I = moment of inertia of cross section of beam with respect to neutral axis

S = stress expressed in pounds per square inch

$$I = \int y^2 da$$

where y = distance of section from neutral axis

da = element of area

when evaluated from $-\frac{h}{2}$ to $+\frac{h}{2}$

$$I = \frac{1}{12} bh^3$$

where b = width of specimen

h = thickness

Calculated values for S within the elastic limit are given in Table I. The equation holds only within the elastic limit.

COMPARISON OF THE EFFECT OF BENDING, ELONGATION BY TENSION,
AND REDUCTION BY COLD ROLLING UPON THE LAUE PATTERNS.

Preparation of specimens—Specimens of aluminum containing crystals of large size (0.5-1.0 square centimeters in area) and specimens composed of small crystals were prepared. The specimens were first machined to the dimensions desired. The large crystals were prepared by a strain and annealing procedure recommended by Carpenter and Elam (16). The samples were annealed in an electric muffle furnace for 48 hours at a temperature of 400 degrees Cent. (750 degrees Fahr.). They were then allowed to cool slowly to room temperature by lowering the temperature about 25 degrees every 2 hours. These specimens were then strained in tension until they were given permanent set of approximately 2 per cent. The set was varied slightly about 2 per cent. After this strain treatment, the specimens were again annealed. This second anneal was begun at a temperature of 300 degrees Cent. (570 degrees Fahr.), and the temperature gradually raised, over a period of 4 days, until the temperature reached 500 degrees Cent. (930 degrees Fahr.). The temperature was then raised to 560 degrees Cent. (1040 degrees Fahr.), for one hour. The final step was to lower the temperature slowly over a period of 2 or more days until the specimens reached room temperature. The specimens were made from commercially pure aluminum. The polycrystalline samples were prepared by merely giving the aluminum strips the first anneal in the procedure for preparing large crystal material. The dimensions of the specimens were as follows: thickness = 0.018 inches, width = 0.300 inches, length was variable.

Cold working—Both polycrystalline and single crystal specimens were subjected to cold working by bending, rolling and elongation by tension.

The specimens were bent by bending them about half cylinders cut from cylindrical tubing. The specimen was held snugly against the tubing by screws placed in the ends of the specimen. A small hole was drilled in the tube through which the X-ray beam passed. The specimen held in its circular-bent position was mounted on a support and a Laue pattern taken. The specimen was then placed on smaller cylinders.

The rolling was done by a rolling machine in which the diameter of the rolls was 2 inches. The specimens tended to curl slightly.

Any curling was removed by subjecting the specimen, held between wooden blocks, to slight pressure in a vise. The data on rolling are given in Tables III and IV.

Elongation by tension was produced by a machine designed by Clark and Beckwith (1) and used for producing tension in metal

Table II
Single Crystal-Aluminum Elongated by Tension

Experiment	Fig.	Stress lbs./in. ²	Per Cent Elongation	γn	$\Delta \delta$ Elongation of Laue Spot-Degrees
AS2	24	0	0	0	0.50 (zero value)
AS3	..	2830	0.418	0.00566	0.5°
AS4	26	3680	1.29	0.0184	0.7°
AS5	..	4700	2.38	0.03225	1.8°
AS6	..	5380	3.12	0.0433	2.1°
AS7	..	6000	3.90	0.0555	2.65°
AS8	..	6170	4.89	0.0690	3.35°
AS9	..	6500	6.23	0.0863	4.5°
AS10	..	6590	7.68	0.1034	5.26°
AS11	28	7020	8.69	0.1175	5.80°
AS12	30	7270	9.97	0.1350	7.80°

Table III
Rolled Single Crystal-Aluminum

Experiment	Fig.	Per Cent Reduction	γn	$\Delta \delta$ Elongation of Laue Spot-Degrees
RoS2	23	0	0	0.40 (zero value)
RoS3	..	0.847	0.01313	0.1
RoS4	25	1.37	0.0230	1.60
RoS4 ¹	..	4.05	0.067	4.42
RoS6 ¹	27	6.85	0.1200	6.82
RoS5	29	7.28	0.1238	6.75
RoS8	..	41.5	0.875	could not measure

specimens. The stress was measured by use of a calibrated spring and an Ames dial. A traveling microscope was used to measure accurately the distance between two fine lines on the specimen. From these measurements, the strain was calculated. The strain and stress are recorded in Tables II and VI.

Registration of Laue patterns—The usual technique for registration of Laue patterns was used. Specimen to film distance was 4 centimeters. The rolled and bent specimens were supported in a constant position. Tension specimens were held firmly in the tension machine throughout the work on that specimen. The tension machine was small enough to fit on the table of the X-ray assembly between the defining pinhole system and the film. Polychromatic copper radiation was used with the tube operating at 20 milliamperes and 30

Table IV
Bending of Single Crystal-Aluminum

Experiment	Fig.	Curvature in Reciprocal cms.	$\Delta \delta$ Elongation of Laue Spot-Degrees
AB1	16	0	0.40 (zero reading)
AB2	17	0.351	1.5°
AB3	..	0.550	2.4°
AB4	..	0.690	3.20°
AB4 ¹	18	1.18	5.0°
AB5	19	1.64	7.0°

Table V
Rolled Polycrystalline-Aluminum

Experiment	Figure	γn
RoP1	32	0
RoP2	34	0.00165
RoP3	36	0.1485
RoP4	38	0.2825

kilovolts. Under these conditions, the time of exposure was 30 minutes.

Orientation of crystals—The orientations of the crystals were again determined by the use of Majima and Togino's patterns. The orientations of the crystals and the Laue spots used for measuring rotation were as follows:

Specimen used for rolling

$$\alpha = 15^\circ$$

$$\beta = 5^\circ$$

Laue spots used = 220, 240

Specimen used for tension

$$\alpha = 35^\circ$$

$$\beta = 20^\circ$$

Laue spots used = $13\bar{1}$, $2\bar{2}0$

Specimen used for bending

$$\alpha = 30^\circ$$

$$\beta = 30^\circ$$

Laue spots used = $20\bar{2}$, $11\bar{1}$

where: α = angle 100 plane makes with vertical plane of X-ray beam.
 β = angle 100 plane makes with horizontal plane of X-ray beam.

Measurement of elongation of Laue spot—The elongation of the Laue spots for the single crystals were measured in degrees for all three types of cold working. Measurements were made as described for the silicon steel. Two spots were selected for measurements in each case and the average value taken. A traveling microscope was used for making measurements on the film. Elongations of the Laue spots are plotted against bending curvature in Fig. 14.

Calculation of amount of cold work—Nadai (17) has set up a mathematical theory of plastic strain which can be used for expressing plastic strains of finite magnitude in terms of the octahedral shear. He has developed equations from which the octahedral shear may be computed from the specific shear or the principal specific ex-

Table VI
Polycrystalline Aluminum Elongated by Tension

Experiment	Figure	Stress lbs./in. ²	Per Cent Elongation	γ_n
TP1	32	0	0	0
TP2	34	2880	0.38	0.0054
TP10	36	9210	12.2	0.1604
TP11	38	9700	19.45	0.252

tensions or contractions. These equations were used for calculation of the octahedral shear of the rolled and elongated specimens. By expressing the amount of cold working in terms of the octahedral shear for each type of cold working, the effect of each type of cold working upon the Laue patterns could be compared. Unfortunately, no equations have yet been worked out for bending. In the case of the crystal elongated by tension, the octahedral shear was calculated from the equation.

$$\text{Octahedral shear} = \gamma_n = \sqrt{2} \ln (1 + \epsilon_1)$$

$\epsilon_1 = \text{strain}$

For simple shear, as in rolling wide sheets

$$\text{Octahedral shear} = \gamma_n = -2(2/3)^{1/2} \ln (1-g)$$

$g = \text{reduction of thickness}$

The octahedral shear was plotted against elongation of the Laue spot for the rolled and elongated single crystal-aluminum in Fig. 15.

DISCUSSION OF RESULTS

Discussion of Results of Experiments with Silicon Steel—Some of the X-ray diffraction patterns of the single crystal of silicon steel registered under the conditions of Table I are shown in Figs. 2-11, inclusive. The average elongation of the two Laue spots used are plotted against the curvature of the specimen in Fig. 13.

Fig. 2 is the Laue pattern of the unstrained crystal. As the crystal is bent, the Laue spots begin to elongate as shown in Fig. 3. When the stress was removed in all of the experiments up to experi-

ment D21, the crystal returned to its normal position and the Laue pattern was indistinguishable from that of Fig. 2. Beginning with experiment D21, the specimen refused to return to its normal position upon the removal of the stress and the Laue pattern showed residual elongation as shown in Fig. 5. Up to D21 all of the rotations or bending of the reflecting planes was elastic. This work within the

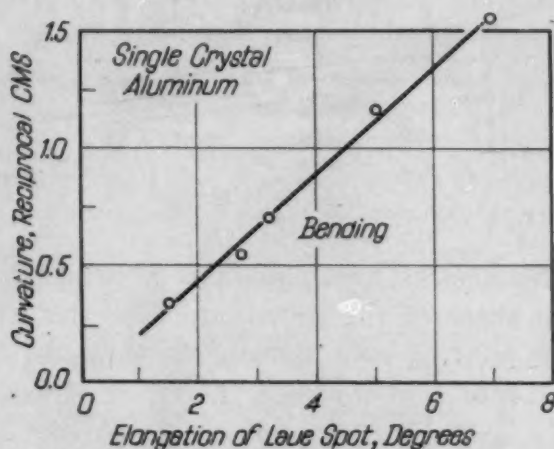


Fig. 14—Elongation of the Laue Spot as a Function of Bending Curvature for a Single Crystal of Aluminum.

elastic limit shows definitely that asterism (elongation of Laue spots) does take place within the elastic limit but disappears upon the removal of the distorting forces. This is in disagreement with the findings of Cox and Blackhurst (18) who failed to find any asterism in a tungsten wire stretched elastically. Clark (19) detected traces of asterism in elastically stretched polycrystalline silicon steel. On the other hand, Nusbaum (20) states that in a series of stress experiments (both tension and bending) on single and multicrystalline materials, he never observed "radial asterism" present until the sample under stress had definitely passed the elastic limit of the material. He states that plastic flow is a necessary condition for the presence of "radial asterism." In spite of what Nusbaum states, experiences with other elastically bent specimens in this laboratory, such as mica, has shown that marked asterism can occur well within the elastic limit.

For specimens bent severely, as shown in Fig. 10, the Laue spots were found to elongate along the curves calculated by Leonhardt. Fig. 12 is one of Leonhardt's curves upon which has been superimposed the Laue pattern of a severely bent crystal of silicon steel.

It is obvious from an inspection of this figure that the Laue spots tend to elongate along the continuous lines of Leonhardt's curves. In this diagram, the bending axis is perpendicular to the line marked 90 degrees and in the plane of the diagram. The continuous lines are lines of constant δ values, where δ is the term used and defined in equation 1.

Fig. 13 is a plot of the elongation of the Laue spot in degrees

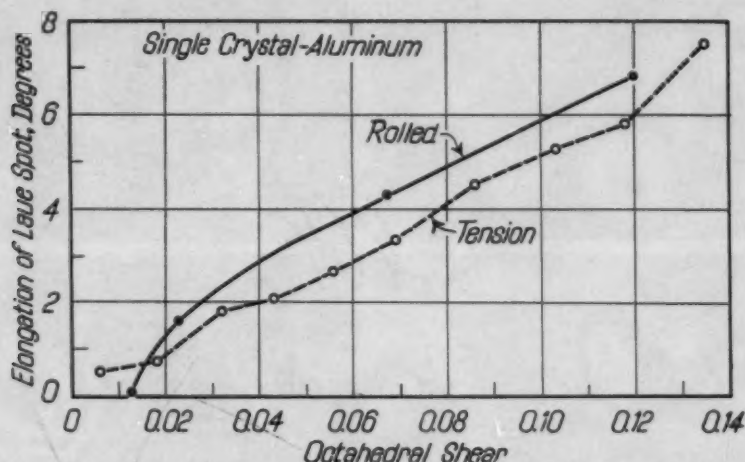
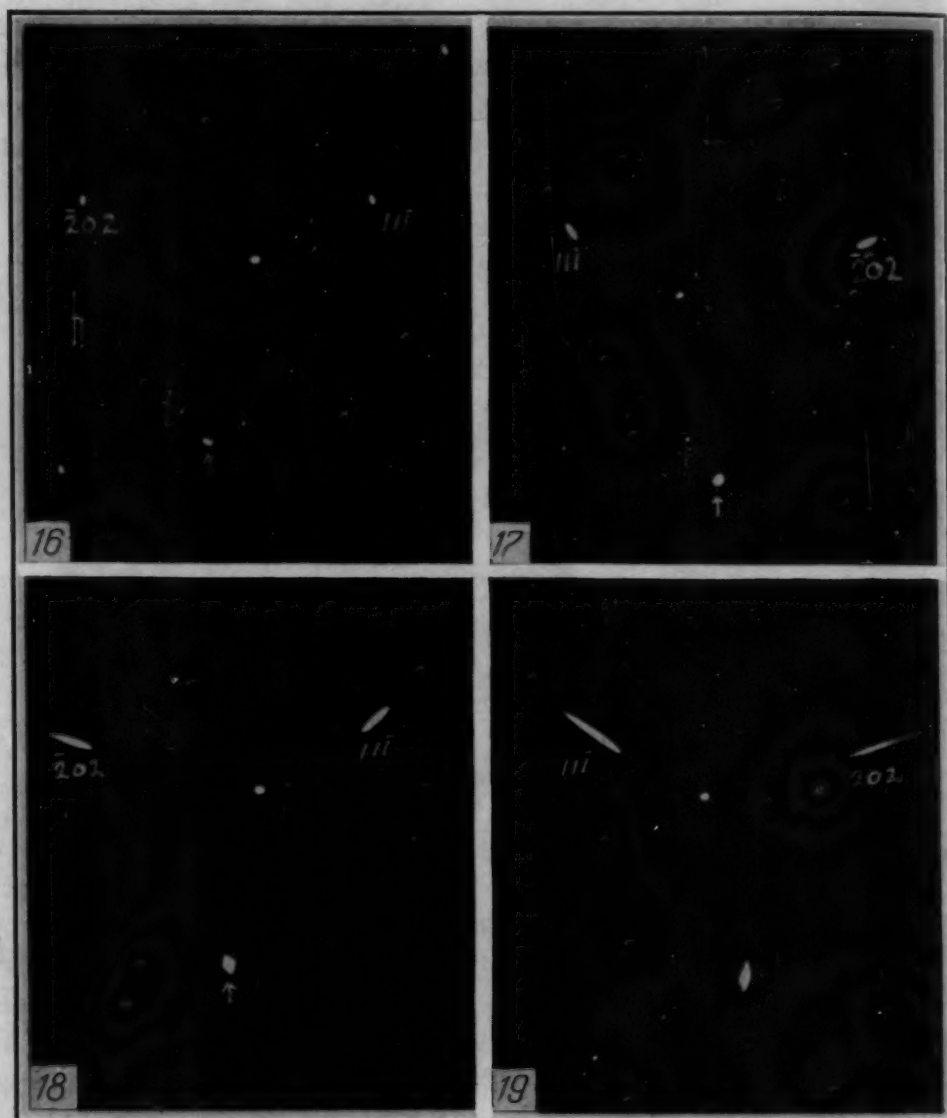


Fig. 15—Octahedral Shear as Related to Elongation of Laue Spots for Deformation of Single Crystal of Aluminum by Rolling and by Tension.

against the curvature. The curve marked "forced" is for the specimen under applied stress. When the stress was removed, some of the elongation of the Laue spot disappeared. The amount of residual elongation of the spots remaining after the bending force is removed is shown in the curve marked "residual." These two curves show that in the range of stresses studied a certain fairly constant amount of the bending or rotation of the planes responsible for the spots is elastic.

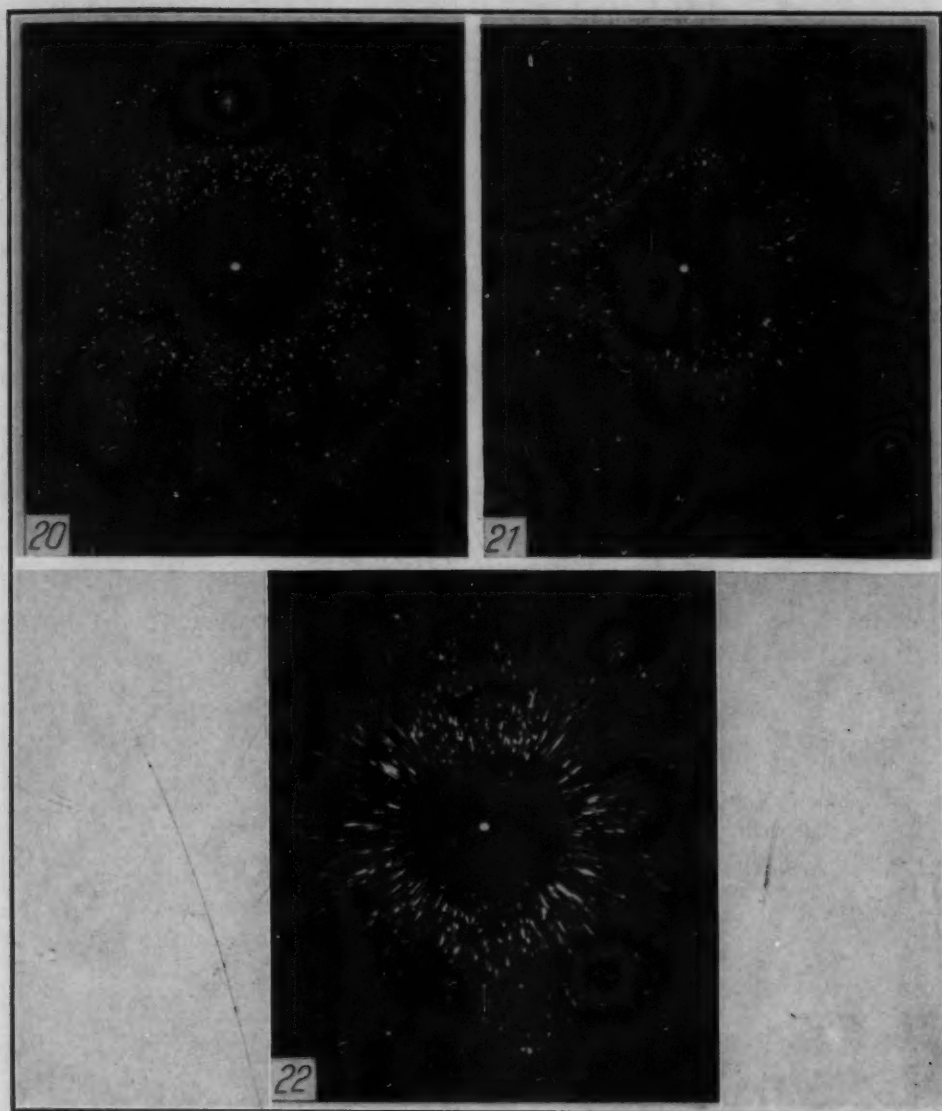
The break in the "forced" curve which appears when the Laue spots have been elongated about two degrees corresponds to the elastic limit of the specimen.

Discussion of results of bending single and polycrystalline aluminum—The Laue patterns of the single crystal aluminum subjected to measured amounts of bending are shown in Figs. 16-19, inclusive. Table IV gives the amounts of bending and the corresponding elongation of the Laue spots for the single crystals. Fig. 14 is a plot of the curvature of the specimen versus the average elongation of



Figs. 16 to 19—Laue Patterns of Single Aluminum Crystal Subjected to Bending (Table IV).

the $\bar{2}02$ and $11\bar{1}$ spots. It is observed that the curve is a straight line. All this bending is well beyond the elastic limit. These specimens were bent about an axis which was in the plane of the patterns and parallel to the longer dimension. The spots, in general, elongated along Leonhardt's curves shown in Fig. 12. It is observed that the three dominant central spots vary in length from one to the other. This is due to their different positions. The length of the Laue spot is a function of δ and σ ; σ as defined previously is roughly equal to the angle between the bending axis and a line drawn from the Laue



Figs. 20 to 22—Patterns for Polycrystalline Aluminum Subjected to Increasing Bending Stress.

spot. As σ approaches 90 degrees, the spot approaches a maximum in length. One of the three dominant spots (the lower one) elongated very slightly and exhibited some peripheral broadening. This is as it should be since σ is nearly zero for this spot.

Some of the patterns of the polycrystalline aluminum specimens subjected to bending are shown in Figs. 20-22, inclusive. The orientation of the bending axis is the same as for the single crystal samples. No attempt was made to measure quantitatively the effect of the bending upon the patterns for the polycrystalline aluminum. Qualita-

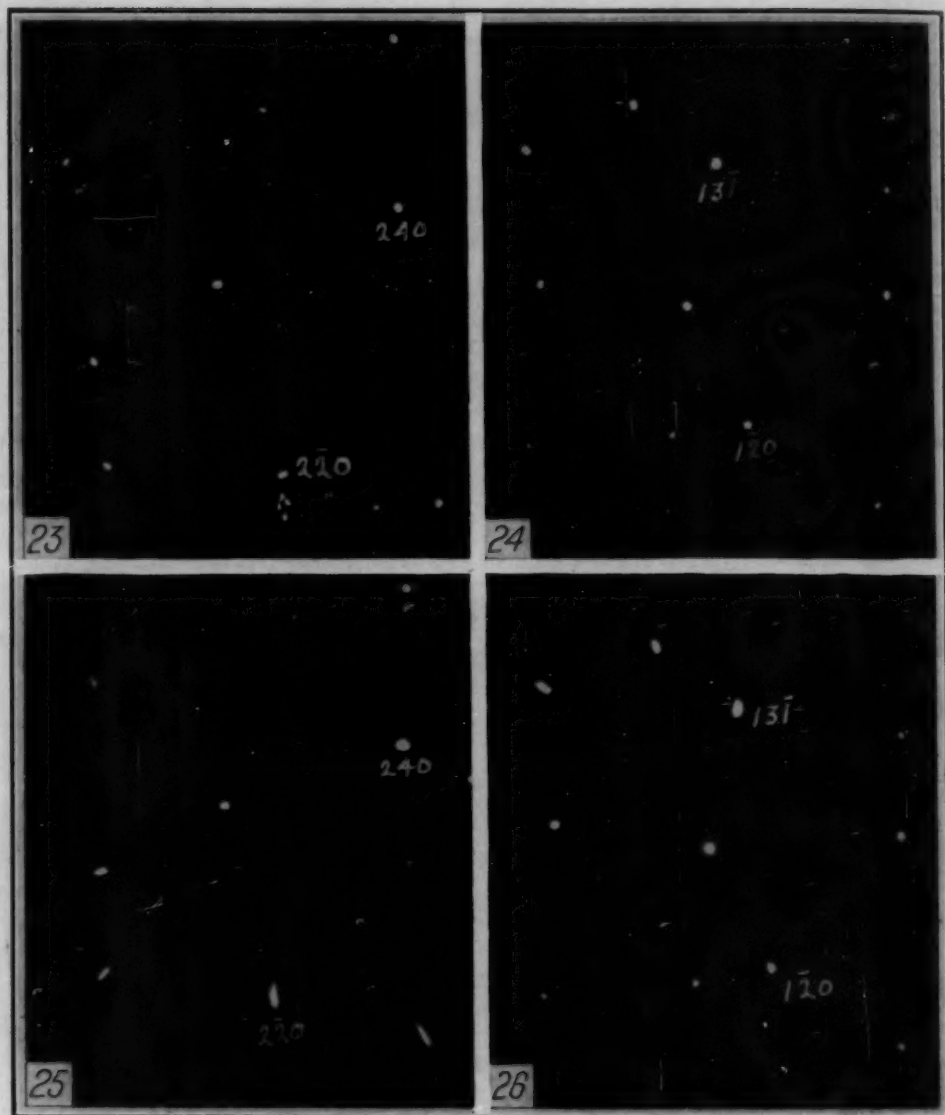


Fig. 23—Single Aluminum Crystal, Original, Before Rolling.
Fig. 24—Single Aluminum Crystal, Original, Before Elongation by Tension.
Fig. 25—Crystal in Fig. 23, Rolled to Reduction of 1.37 Per Cent.
Fig. 26—Crystal in Fig. 24, Elongated 1.29 Per Cent in Tension.

tively, the effect was like that for single crystals. The spots were elongated as the specimens were bent, and close observation reveals that the elongation is greater as σ approaches 90 degrees. The spots along a line through the central spot and parallel to the longer dimensions of the patterns appear less elongated than the others. These are the spots for which $\sigma = \text{zero}$. Due to the complex nature of the stresses acting upon a small crystal in an aggregate as in the case of polycrystalline aluminum, the spots do not follow Leonhardt's curves so well.

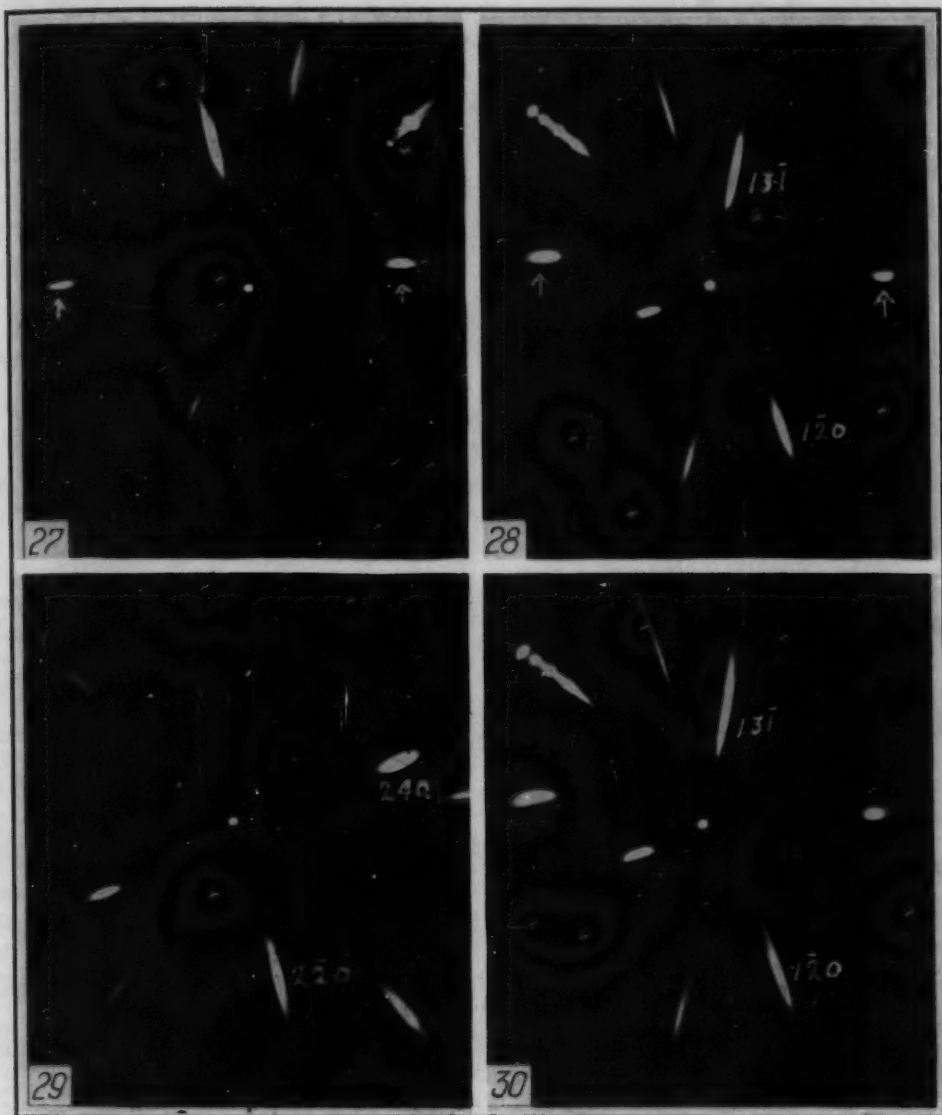


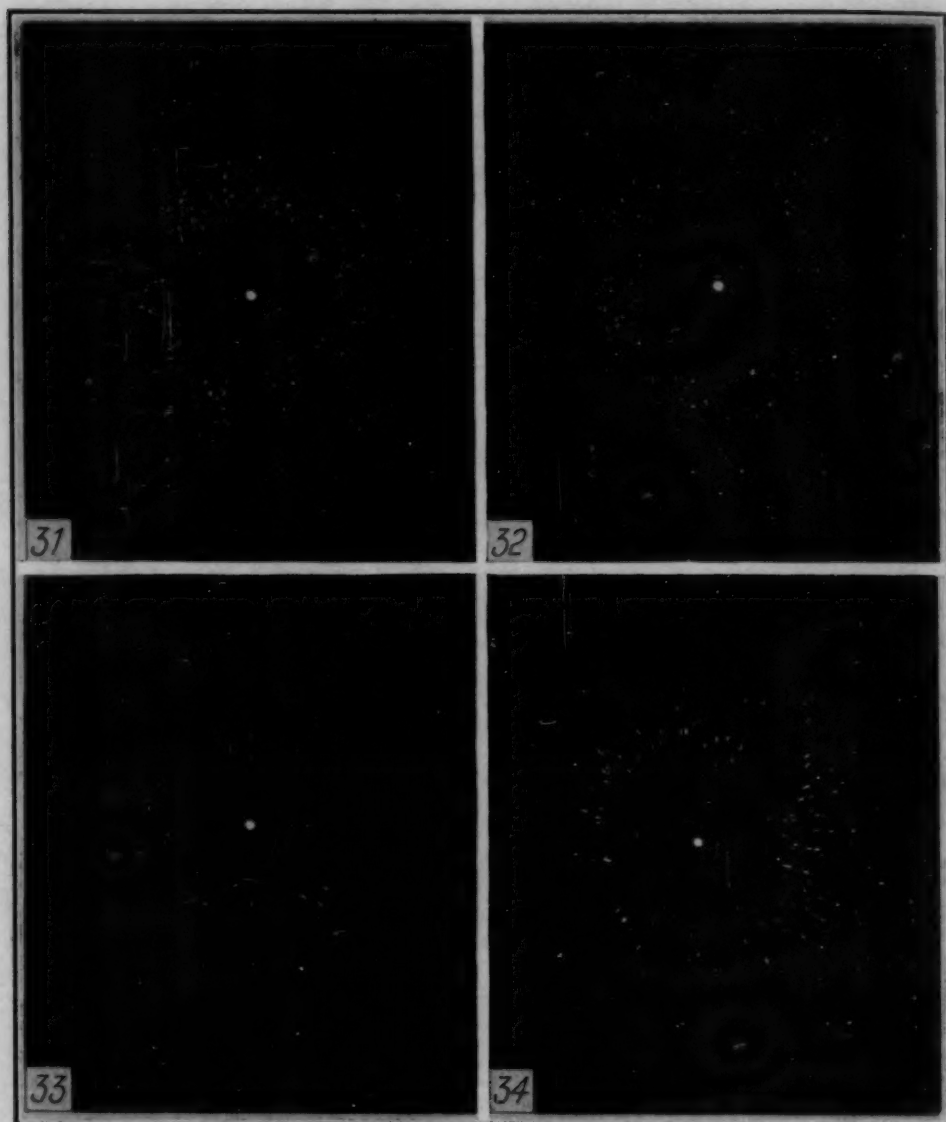
Fig. 27—Crystal in Fig. 23 Rolled to Reduction of 6.85 Per Cent.

Fig. 28—Crystal in Fig. 24 Elongated 8.69 Per Cent.

Fig. 29—Crystal in Fig. 23 Rolled to Reduction of 7.28 Per Cent.

Fig. 30—Crystal in Fig. 24 Elongated 9.97 Per Cent.

Discussion of results of experiments on rolling and elongation by tension of single and polycrystalline aluminum—Some of the typical patterns of the single crystal aluminum, rolled and elongated by tension, are shown in Figs. 23-30, inclusive. The two sets for rolling and tension are placed side by side for comparison. The similarity of the patterns is obvious. The specimens were elongated in a direction parallel to the longer dimension of the pattern. Rolling was in the plane of the pattern and in a direction parallel to the long



Figs. 31, 33, 35, 37—Polycrystalline Aluminum (Original Fig. 31) Subjected to Increasing Rolling Deformation (Table V).

Figs. 32, 34, 36, 38—Polycrystalline Aluminum (Original Fig. 32) Subjected to Increasing Elongation by Tension in Steps so That Figs. 34, 36, 38 are Comparable with Figs. 33, 35, 37 Respectively (Table VI).

dimension of the pattern. It is observed that the Laue spots elongate in a manner similar to the spots in the bent specimens. The effect is as though the specimen were bent about an axis perpendicular to the direction of elongation and rolling.

The small amount of elongation displayed by the side spots indicated by arrows in Figs. 27 and 28 indicate that these spots lie near the axis about which the crystal is bending or rotating.

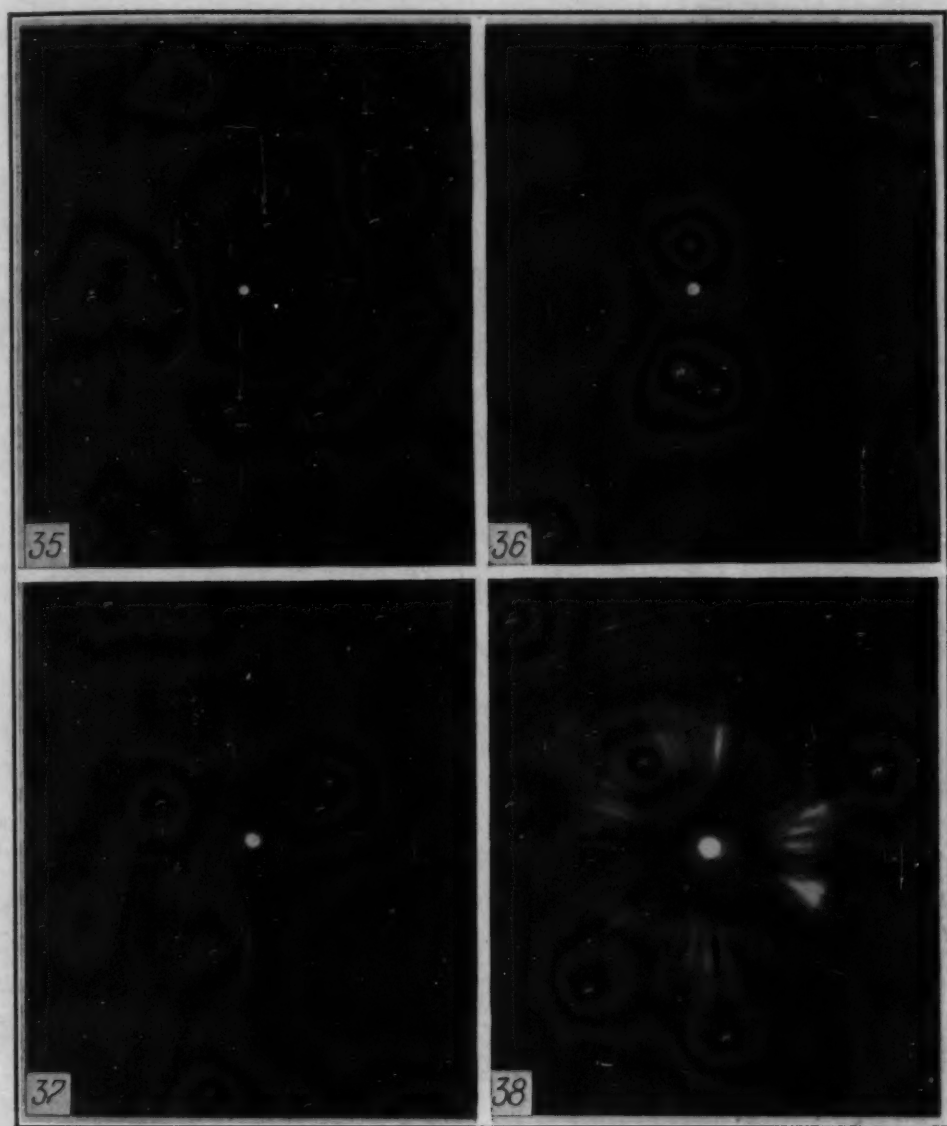


Fig. 15 shows how rolling and elongation when expressed in terms of octahedral shear resemble each other in their effects upon single crystal aluminum. The effect of rolling does not appear as readily as does the effect of elongation. However, the effect of rolling appears suddenly when it does appear and then levels off to a straight line function. Some indication of a periodicity is shown in the curve for tension. A more marked periodicity in deformation was found by Clark and Beckwith (12).

It must be remembered that the cold working here was much less than that required to produce fibering. On the basis of work

that has been done on extreme cold working by elongation and rolling, it seems that the two types of cold working would resemble each other less in their effects in these extreme cases. Differences in orientation seem to appear.

Some typical patterns of polycrystalline aluminum cold worked by rolling and elongation by tension are shown in Figs. 31-38, inclusive. No quantitative work was attempted on these. An inspection of the patterns shows the resemblance in effect of the two types of cold working upon the polycrystalline material.

CONCLUSION

From the work on silicon steel, the conclusion was reached that the curvature of a single crystal of silicon steel may be correlated with the elongation of the Laue spots in degrees as calculated from Leonhardt's equations.

Under the conditions of this investigation "radial asterism" does take place within the elastic limit when single crystals of silicon steel are bent. This opposes the findings of Nusbaum (20). Beyond the elastic limit, a part of the rotation or bending of the planes is elastic. Whether this radial asterism is due to bending of planes or rotation of "blocks" was not decided from this work.

Equivalent amounts of cold working, when expressed in terms of Nadai's octahedral shear, whether by rolling or elongation by tension within the limits of deformation used in this investigation, produce a similar effect as revealed by the X-ray patterns. No quantitative measurements of the type made in this investigation can be made for very severe cold working on metals of the crystal sizes used because of the diffuseness of the Laue spots.

In bending, the planes tend to rotate or bend about the axis of bending as one would expect. The planes in a crystal subjected to rolling tend to rotate or bend about an axis in the plane of rolling and perpendicular to the direction of rolling. Elongation by tension causes the planes to rotate or bend about an axis perpendicular to the direction of elongation.

Nadai's "octahedral shear" used for expressing the amount of cold work may be correlated with the X-ray patterns of the cold-worked metal for the specimens and conditions of this investigation.

References

- 1a. G. L. Clark and M. M. Beckwith, *TRANSACTIONS, American Society for Metals*, Vol. 25, 1937, p. 1207.
- 1b. G. L. Clark and C. G. Dunn, *Phys. Rev.*, Vol. 52, 1937, p. 1170.
2. S. L. Hoyt, *TRANSACTIONS, American Society for Metals*, Vol. 24, 1936, p. 789.
3. H. J. Gough, *Proceedings, American Society for Testing Materials*, Vol. 33, 1933, p. 53.
4. F. Rinne, *Zeit. f. Krist.*, Vol. 59, 1924, p. 230.
5. J. Leonhardt, *Zeit. f. Krist.*, Vol. 61, 1924, p. 100.
6. S. Goucher, *Philosophical Magazine*, Vol. 2, 1926, p. 289.
7. Regler, *Z. Physik*, Vol. 74, 1932, p. 547.
8. *Chemistry and Industry*, Vol. 52, 1933, p. 317.
9. L. Froumer, *Metal Industry (London)*, Vol. 54, 1939, p. 435-40.
10. W. A. Wood, *Philosophical Magazine*, Vol. 1, 1931, p. 610.
11. G. W. Brindley and P. Ridley, *Proceedings, Physical Society (London)*, Vol. 51, 1939, p. 432-48.
12. G. L. Clark and M. M. Beckwith, *Zeit. f. Krist.*, Vol. 90, 1935, p. 392.
13. Jacquesson, *Mecanique*, Vol. 23, 1939, p. 156-62.
14. F. Bollenrath and E. Osswald, *Z. Metallkunde*, Vol. 31, 1939, p. 151-9.
15. M. Majima and S. Togino, *Sci. Papers Inst., Phys. Chem. Res. Tokyo*, Vol. 7, 1927, p. 75.
16. Carpenter and Elam, *Proceedings, Royal Society*, Vol. A100, 1921, p. 329.
17. J. Nadai, *Applied Physics*, Vol. 8, 1937, p. 205-217.
18. Cox and Blackhurst, *Philosophical Magazine*, Vol. 46, 1929, p. 981.
19. G. L. Clark, "Applied X-Rays," p. 513, 3rd Ed., McGraw-Hill Book Co., 1940.
20. C. Nusbaum, *TRANSACTIONS, American Society for Metals*, Vol. 24, 1936, p. 1029.

DISCUSSION

Written Discussion: By C. G. Dunn, Magnetic Section, General Electric Co., Pittsfield, Mass.

Radial asterism obtained from single crystals or coarse-grained samples clearly indicates a bent condition in the reflecting planes. The mosaic structure of crystals probably affects the details of the bending, but does not prevent an elastic bending of the reflecting planes. Asterism alone would not differentiate between an elastically and an inelastically bent plane; but other data can, and these the authors have obtained it. The authors are also to be commended for their correlation work showing a linear relationship between the curvature of the specimen and the elongation of the Laue spots.

The curvature of that part of the specimen in the path of the X-ray beam should, of course, bear a linear relationship to the angular elongation of the Laue spots. However, there is some question as to the magnitude of this elongation. Simple calculations based on a pin-hole of 0.025-inch diameter indicate that a curvature of 0.191 cm.^{-1} for instance, (see Table I) should bend a crystal by only 0.70 degree, but the data show a considerably larger value.

It is easy to see that asterism should be obtainable by elastic bending of crystals because of the following factors:

The maximum per cent strain in a specimen of thickness t bent into an

arc of radius r is $100t/2r$; so for a given r of reasonable value a t can be found to fall within the elastic range. The above example shows the per cent strain to be $100 \times 0.018 \times 0.191/2$ or 0.17 per cent and Table I shows this to be within the elastic limit. By making t smaller it is possible to get a still larger curvature within the elastic limit.

Written Discussion: By A. Nadai, consulting mechanical engineer, Westinghouse Electric and Manufacturing Co., E. Pittsburgh, Pa.

The authors have carefully investigated the effects of distortion of the X-ray patterns obtained in their single crystal and polycrystalline specimens while these latter were held under stress and after the load was released. It should be noted that while all parts of a specimen which was elongated by pure tension or by rolling are strained uniformly (this is true for a distortion through rolling provided the amounts of reduction are comparatively small) in the case of the bending the strains are proportional to the distance of a point from the neutral axis and the bending stresses in the case of plastic bending are dependent in some more complex manner on this distance. Therefore the effects plotted in Figs. 13 and 14 for the bending tests must contain the integrated contributions to the X-ray pattern of all the elements of the specimen. The writer would be interested to know whether by a further analysis of such X-ray patterns also the distribution of the plastic bending stresses for a beam consisting either of a single crystal or of many small crystal grains could be determined. For polycrystalline materials, engineers have worked out methods which allow to predict to a certain degree the distribution of the bending stresses after plastic bending and thus also the distribution of the residual stresses after a release of the external forces. The authors should be commended for further clearing of these conditions. It would be valuable and of considerable interest to utilize perhaps the X-ray analysis for a direct observation of the residual stresses which must remain in the surface layers of a bar after plastic bending or plastic torsion and thus to verify by means of X-rays the theory of macro-residual stresses.

Authors' Reply

Drs. Dunn and Nadai have given valuable suggestions leading to further experimental tests and even to further possibilities of extending interpretation of the patterns presented in this paper. Both attempts are being made. The general approval of this effort by these competent critics is deeply appreciated, and it serves as a stimulus to take the next step forward in this important but difficult field of using X-ray patterns to evaluate both quantity and mechanism of deformation in metals.

EFFECTS UPON FURNACE REFRACTORIES OF PROTECTIVE GASES FOR HIGH CARBON STEELS

BY JOHN H. LOUX

Abstract

Protective atmosphere gas for high carbon steels presents the same firebrick disintegration problem to heat treating furnace engineers that is found in the blast furnace. Blast furnace operators licked their problem many years ago by using a dense surface brick which is given a reducing firing period.

On a recent installation of hood type furnaces for annealing high carbon steels, extensive tests were run to prove that what was occurring in the blast furnace could occur in exactly the same manner in a heat treating furnace. A nondisintegrating type of firebrick was used in the hood type furnace bases and among its advantages over light weight brick were its resistance to abrasion and its heat storage capacity.

Recent work has shown that the reducing firing period must continue to a point beyond which all iron oxides in the brick are changed to the magnetic type. This chemical change is reversible and it is necessary to continue the reducing fire until all iron is absorbed into a silicate glass.

STATEMENT OF PROBLEM

FOR a great many years, operating and ceramic engineers have been faced with the problem of brick disintegration caused by the reaction of certain gaseous atmospheres upon firebrick materials. The purpose of this paper is to review and bring up to date the results of certain conclusive experiments made in this field.

Ten years ago, when special protective gases were quite new in the furnace industry, it was promptly discovered that protective gases for high carbon steels would disintegrate firebrick, while most gases used for low carbon steels would not. It was further discovered that the gases that disintegrated firebrick would not affect a light refractory brick. Instead of finding out why this occurred, the majority

A paper presented before the Twenty-second Annual Convention of the Society held in Cleveland, October 21 to 25, 1940. The author, John H. Loux, is development engineer, Salem Engineering Co., Salem, O. Manuscript received June 11, 1940.

of furnace designers and users simply accepted it as a fact. Whenever a special atmosphere furnace was purchased, it was unanimously understood that all of the refractory within the furnace would have to be of the semi-firebrick class. Furnaces which normally would utilize firebrick piers were built with light refractory. Extreme care was taken to design pier caps and install them in a manner which would prevent abrasion of the softer brick.

PRACTICAL APPLICATIONS

This practice was carried out in both electric and fuel fired radiant tube type furnaces, and was, no doubt, a large contributing factor in the rapid development and general use of light weight refractory materials.

There are many applications where the higher insulating value of this brick is advantageous. This is particularly so if a rapid cooling problem is involved, or if the furnace is being operated intermittently. In contrast to this, however, there are numerous applications where the heat storage capacity of the heavier firebrick is of extreme value and importance.

As an example, consider a hood type furnace for heating a heavily loaded rack of material. The bottom of such a load is always the last point to come to temperature. If the cycle is supposed to be short, and the load can be placed on a hot base, the firebrick is of real assistance in bringing the bottom of the load up to temperature quickly. Although with a required slow heating-up cycle, this advantage is not so pronounced, it is, nevertheless, present.

In the case of the pusher type, roller rail furnaces, using a specially prepared protective gas atmosphere for high carbon steels, there has never been any logical reason for making the supporting piers out of fragile light refractory material except that it was thought impossible to obtain a firebrick that would prove satisfactory and not disintegrate in the gaseous atmosphere.

There are many other applications where firebrick almost invariably has an advantage over the light weight brick. In any location where resistance to abrasion, heat storage capacity, or plain brute strength overshadow all other requirements to be met, firebrick has a pronounced advantage that should not be overlooked.

In direct contrast to the builders of heat treating furnaces, the builders and users of blast furnaces have approached this same prob-

lem of brick disintegration from an absolutely opposite angle. Blast furnace linings have, for many years, disintegrated in the lower temperature ranges of the wall, while the furnace is operating. The disintegration usually occurs just above the mantle in the outer half of the wall and tapers upward working in slowly toward the inner face.

Large cracks appear in the brick, accompanied by disintegration and finally crumbling. It has been definitely established that the disintegration takes place in a certain low temperature range and does not operate at higher or lower temperatures.

Furnas,¹ in his abstract before the American Ceramic Society in 1936 says:

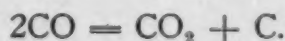
"Careful examination reveals that disintegration is always accompanied by localized deposits of carbon. These deposits are in the form of amorphous carbon and are in small spherical globules, usually about the size of a pea, but sometimes they are as much as three-quarters of an inch in diameter.

"These carbon deposits are magnetic, proving the presence of iron or magnetic iron oxide, or both. Analysis shows that the globules contain 70 to 90 per cent carbon.

"Large cracks always radiate from a carbon segregation as soon as it has arrived at a reasonable size, and apparently it is here that the large cracks, which are characteristic of disintegration, start.

"A small volume of iron oxide tested in an atmosphere of carbon monoxide at 450 to 500 degrees Cent., will deposit a really large volume of carbon. Nesbit and Bell² have made some quantitative determinations of this effect and have found that ferric oxide will deposit as much as 70 times its volume of carbon.

"It would seem then that in the course of disintegration a great deal of carbon must result from the reaction



"Most of the carbon is deposited according to the equation but there must be some agency to urge it on for it does not occur under ordinary conditions. Something in the nucleus started by the deposit of carbon by the oxide acts as a catalytic agent and causes the carbon monoxide to begin adjusting itself to equilibrium condition with approximately 3 per cent carbon monoxide in the gaseous phase. It

¹C. C. Furnas, "Kinetics of Some Reactions of Interest to Ceramists," *Journal, American Ceramic Society*, Vol. 19, No. 6, June 1936.

²C. E. Nesbit and M. L. Bell, "Firebrick Disintegration," *Brick Clay Record*, Vol. 62, 1923, p. 1042-43.

may be some peculiarity of the oxide present, or it may be that carbon once started to deposit forms a crystalline structure which tends to grow even against enormous pressure, just as a crystal of salt in a supersaturated solution will grow very rapidly. In all events, it seems almost certain that ferric oxide starts carbon depositing and that such carbon will deposit indefinitely even after it has burst the brick."

Dr. J. S. Unger,³ of Pittsburgh, who did some later work states:

"It has been found that magnetic oxide alone will not produce disintegration; neither will the nonmagnetic. It must be accompanied by a small quantity of metallic iron to get the proper catalyst. The thing that makes the brick break up or disintegrate is a mixture of ferric oxide and metallic iron."

If blast furnace operators had followed the line of least resistance when faced with the problem of disintegration and substituted light weight refractories they would have failed utterly. To date, at least, no light weight brick has been developed that will stand the abrasion and abuse to which the blast furnace lining is subjected. Instead of running from their problem, blast furnace builders have found out why the problem exists and, furthermore, what they can do about it.

PREVIOUS WORK

Numerous tests have been run by those interested in blast furnace brick. Top brick 40 per cent plastic, 60 per cent flint; wall brick 25 per cent plastic and 75 per cent flint; hearth and bosh 15 per cent plastic and 85 per cent flint, have all been disintegrated in laboratory CO tests within 12 hours.

Usually, the laboratory gas contained from 60 to 95 per cent of carbon monoxide and the temperature ranged from 450 to 500 degrees Cent. (840 to 930 degrees Fahr.). In many samples large pieces of carbon could be plainly seen on the face of the brick or on the disintegrated portions.

The types of brick tested covered a wide variety of makes and, strangely enough, although some disintegrated rapidly, others would not disintegrate at all. At first it was thought that inasmuch as iron

³Dr. J. S. Unger, Discussion of Paper by Roy A. Lindgren, "Some Observations Regarding Refractories for Iron Blast Furnaces," Cleveland Meeting American Institute of Mining and Metallurgical Engineers, October 1936. *Transactions*, Vol. 125, Iron and Steel Division, 1937, p. 88-90.

oxide is an important constituent in causing the reaction, that perhaps the greater the amount of iron oxide content, the greater the disintegration rate. One analysis, however, proved that the disintegrated brick contained 1.71 per cent iron oxide, while the nondisintegrated brick contained 2.28 per cent iron oxide.

Further tests and analyses brought out the fact that the iron oxide segregations in the inert brick were of the magnetic type, while most of those in the disintegrated brick were of the nonmagnetic type (Fe_2O_3).

Again quoting Mr. Furnas:

"The following theory has been formulated to explain the difference in the actions of these two kinds of brick.

"The brick are made from very much the same kind of clay and in both cases the iron occurs in much the same form. The iron may occur in the clays in the form of any of its oxides, as iron pyrites (FeS_2), or as iron carbonate (FeCO_3).

"In firing brick all of these various forms of iron tend to become Fe_2O_3 , provided a good supply of air is kept up during the firing. FeS_2 decomposes to give Fe_2O_3 and SO_2 . FeCO_3 decomposes to give CO_2 and FeO , which, in turn, is oxidized to Fe_2O_3 . Very little, if any, FeO occurs free in the clays. Analysis for this constituent failed to show more than a trace. Carbonaceous matter in brick stops oxidization of iron and may even reduce the iron present. If a strongly oxidizing atmosphere is not maintained in the kiln during firing the iron will end up in a partly reduced condition. In the drying and baking, the nondisintegrated brick must have been subjected to a reducing treatment which produced the magnetic oxide (Fe_3O_4). The finished nondisintegrated brick then contained this magnetic oxide which will not start a nucleus of carbon to burst the brick. The treatment for the disintegrated brick must have been such that the oxide nearly all finally occurs as Fe_2O_3 and so is susceptible to disintegration."

To prove his theory, Mr. Furnas made up a load of brick from a dry mix of 75 per cent flint and 25 per cent plastic clays. These clays were mixed with oxides to produce a resultant magnetic oxide (Fe_3O_4) at from 3.44 to 3.70 per cent and an oxide (Fe_2O_3) from 3.37 to 3.63 per cent. Half of these bricks were fired in an oxidizing atmosphere and half of them in a reducing atmosphere. The firing period was 96 hours. The maximum temperature of the reducing atmosphere was 1350 degrees Cent. (2460 degrees Fahr.), and the

maximum temperature of the oxidizing atmosphere was 1325 degrees Cent. (2415 degrees Fahr.). These bricks were then subjected to a disintegrating test and the results were exactly as predicted. All of those fired in the reducing atmosphere stood the test, while those fired in the oxidizing atmosphere only disintegrated rapidly.

Dr. Unger carries this explanation still further by stating that in addition to converting the iron oxide into a magnetic state this reducing period compels the metallic iron present to combine with the free silica in the brick to form a ferrous silicate, which is inert.

Although for many years in brick-making circles, the thoroughly oxidized brick has been rated the best because it is free from sulphur which is supposed to weaken the brick, a great many manufacturers include a reducing firing period in the manufacture of all of their blast furnace brick. They seem to feel that it is better to obtain the virtues of a nondisintegrating product, even if they must sacrifice some other qualities which the strictly oxidizing firing might produce. They could manufacture brick both ways, but they claim it would complicate their storage and delivery problem to such an extent that it would not be desirable.

MORE RECENT THEORY AND RESULTS

In a private communication from S. M. Swain,⁴ we find that more recent work indicates that the reducing firing period must be carried to a point beyond which the iron oxide is changed to the magnetic type and the metallic iron compelled to form ferrous silicate. The chemical change from ferric oxide to magnetic is reversible and operating results have indicated very plainly that if the brick is insufficiently fired and later used in an oxidizing atmosphere a large percentage of the magnetic oxide will revert back to the nonmagnetic. The brick is then no better from a disintegration viewpoint than one fired without a reducing burn in the first place.

The surest way to prevent this reaction from reversing is to continue the reducing burn until both the iron oxide segregations and the metallic iron are compelled to react with the silica in the clay to form a silicate glass.

This silicate glass is absolutely inert. The ratio of iron to silica in the glass may vary over a wide range and its theoretical formation is shown in Fig. 1.

⁴Ceramic Engineer, North American Refractories Co., Cleveland.

During a normal oxidizing fire the iron and silica, although in contact with each other, will not react. As soon as the reducing fire is begun the ferric iron oxide changes to a magnetic oxide:



In this state the iron reacts immediately with the silica and the glass begins to form at the point of contact of the segregations. The process of diffusion continues with the iron and silica gradually being absorbed into the ever growing particle of glass. Because there is no

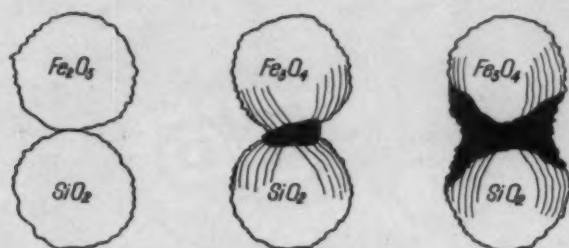


Fig. 1—Theoretical Formation of an Inert Silicate Glass.

definite ratio of iron to silica required to form this glass, its formation continues, regardless of the iron or silica content of the clay.

Eventually, if the reducing burn is continued long enough, all of the iron oxide segregations are absorbed into this glass. Being inert, the glass has absolutely no catalytic properties and will not, under any operating condition, break down into its original constituents.

A very practical method that has been found to check the disintegration resistance of a newly burned brick without resorting to disintegration tests is to check the hardness. Generally speaking, a soft burned brick will disintegrate, whereas a hard burned brick will not.

CORROBORATIVE TESTS

In a recent installation of equipment for the purpose of annealing high chromium steel in bar, tube, and coil form the refractories were subjected to a protective gas atmosphere and were at the same time called upon to bear quite heavy loads.

Hood type furnaces were chosen to do the job with almost the same number of hoods being used as bases. In order to bring the

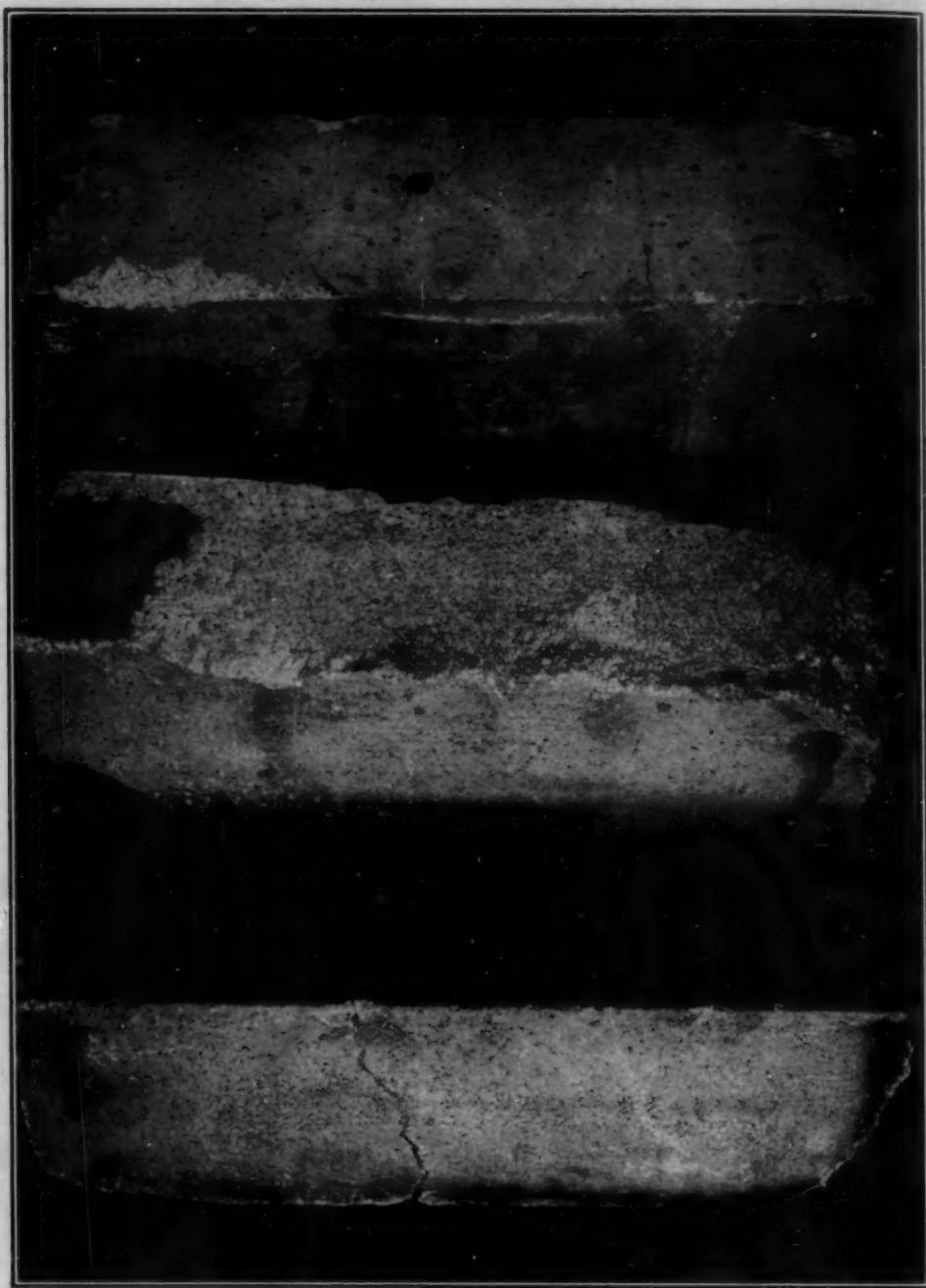


Fig. 2—Results of Exposure to Operating Conditions for Three Months. The Two End Bricks Were Given a Reducing Burn in the Kiln; the Center Brick Was Not.

material out from beneath the inner covers fairly clean and at the same time to prevent all decarburization, a protective gas atmosphere, low in CO_2 but very high in CO was required. This atmosphere was made from natural gas, using a double cracking process.



Fig. 3—The Same Bricks as Shown in Fig. 2. The Center Brick Was Considered Unsatisfactory. The One on the Top Was Almost as Good as When First Placed in Service.

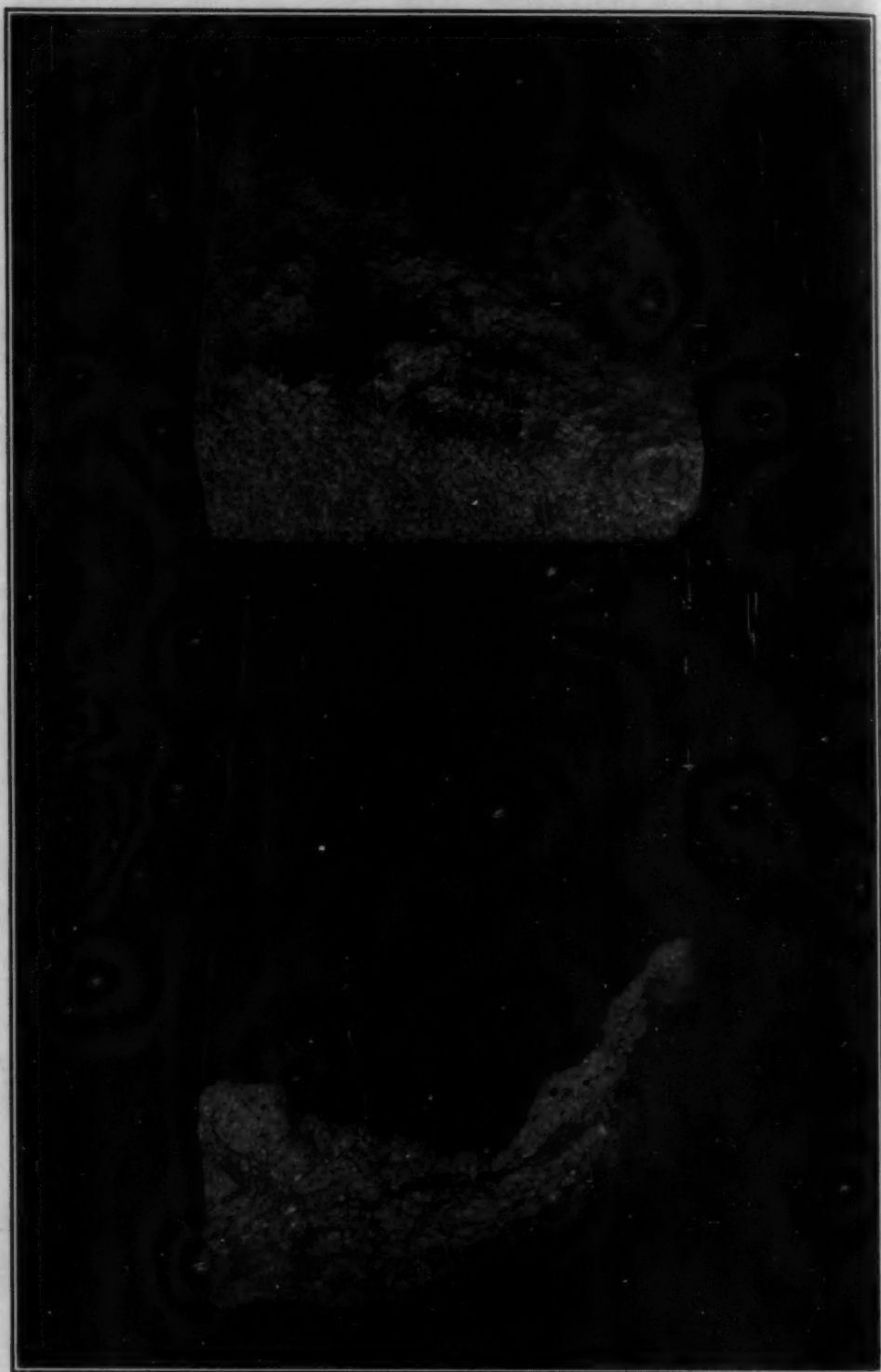


Fig. 4—Most of this Second Quality Firebrick Completely Powdered Away.

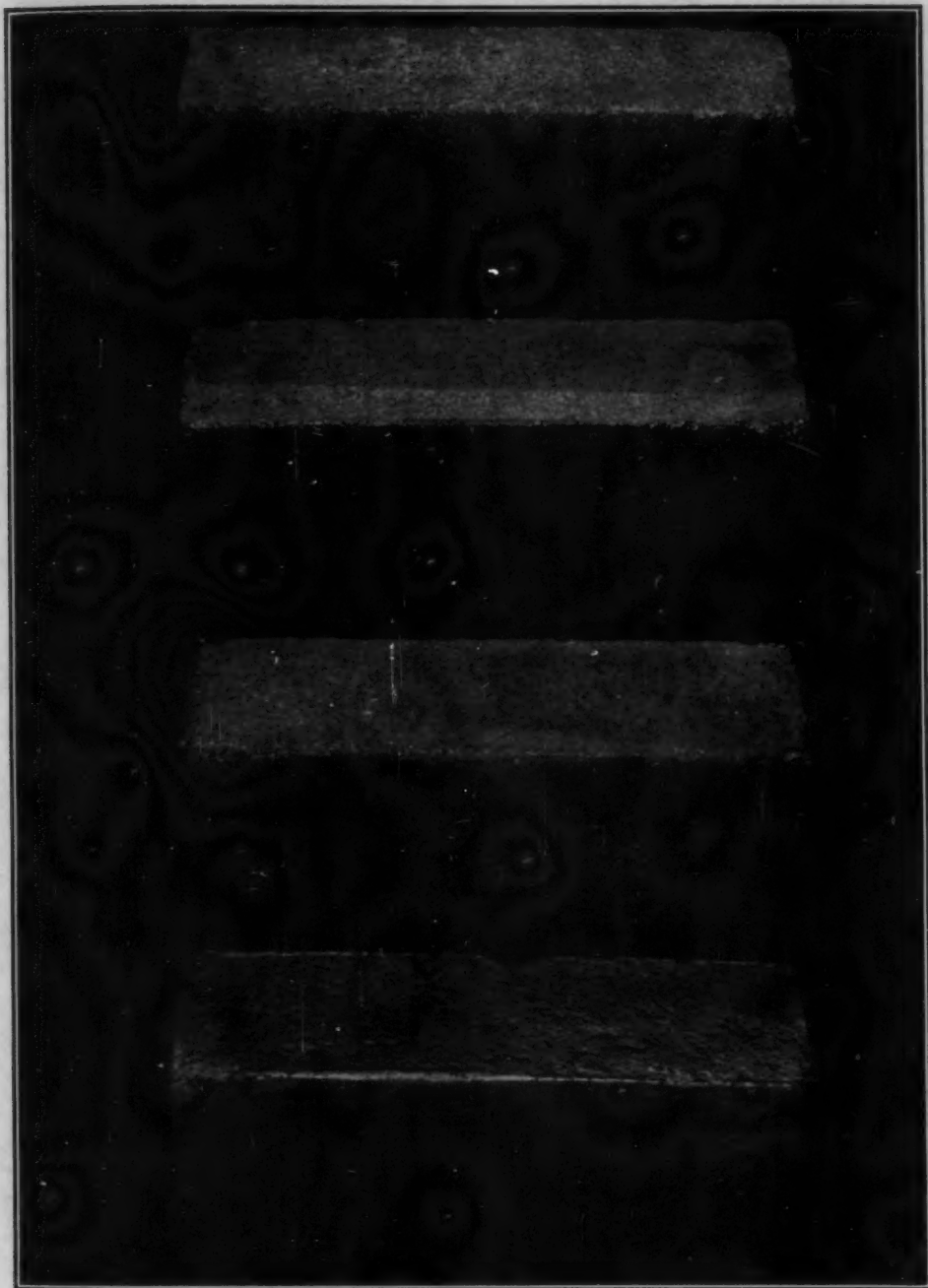


Fig. 5—The Brick on the Bottom is a Light Refractory. The Others are a Cheap Grade of Firebrick and all Were Glazed on Five Sides. No Disintegration Occurred.

This high carbon steel protective gas, therefore, having approximately 20 per cent carbon monoxide content, created exactly the same brick disintegration problem in these special atmosphere furnaces as had, for years, been found in the blast furnace.

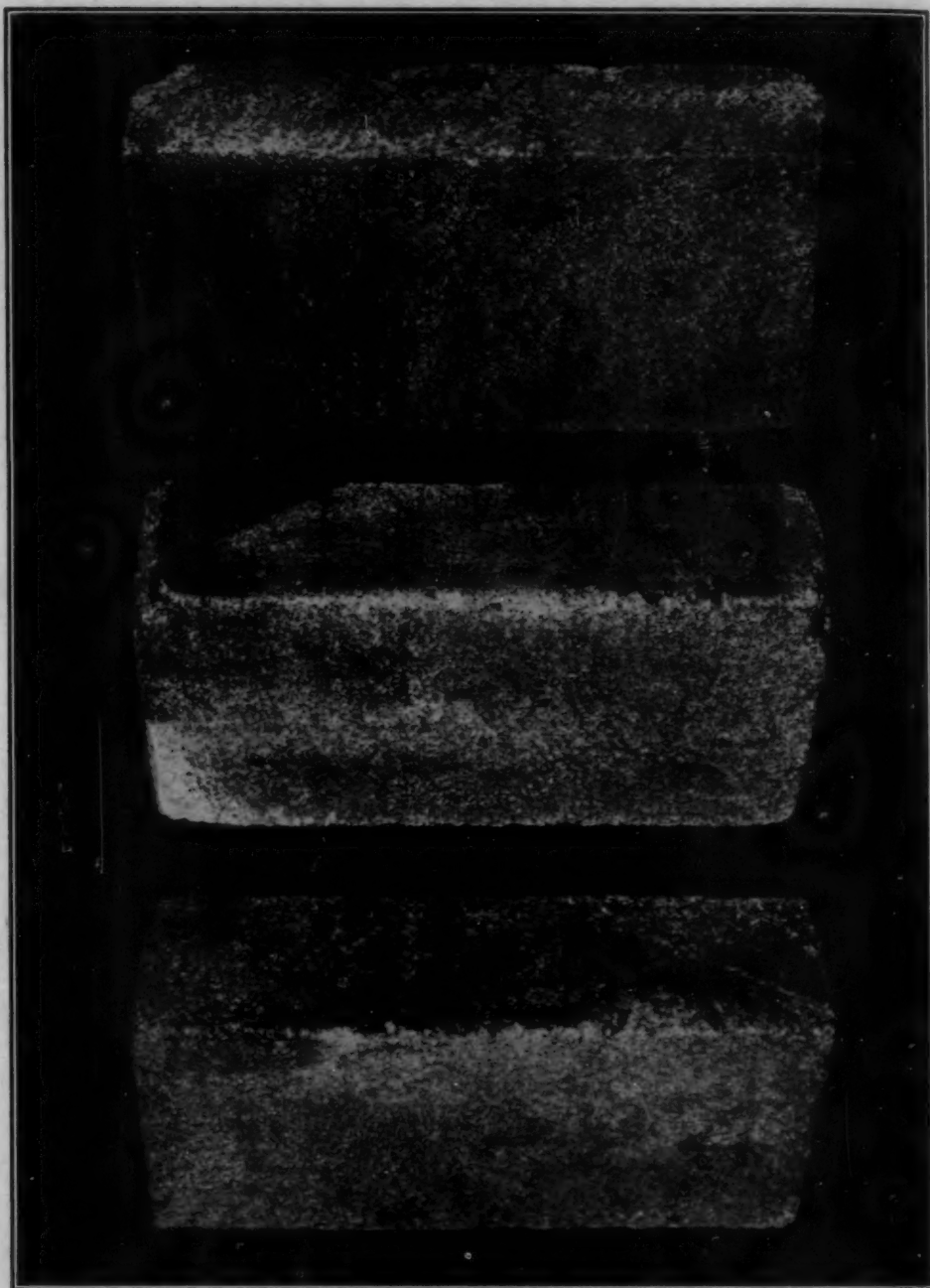


Fig. 6—Except for Mechanical Abrasion All Unglazed Light Refractory Brick Stood the Test.

On the installation of hood type furnaces referred to, a light refractory material was used throughout. This confirms just what has been said about the furnace designers throwing up their hands and accepting the easiest solution to their problem.

A thick plate was laid on top of the refractory bases in order to protect the light brick from abrasion. Stainless steel bars were laid across this plate, and the rack of material placed on the bars.

The results were just what one might expect. After a short time the thick plate took on the appearance of a washboard and the light refractory beneath each stainless steel bar was crushed so that all of the refractory in each base had to be replaced.

Before installing more hoods and bases of this same type this particular company, in conjunction with the furnace builder, ran some rather extensive tests on the various types and grades of brick that might be used.

During these tests, five principal types of brick were built all into one base. These brick were subjected to standard operating conditions and were in no way protected from the temperature or atmosphere gas for a period of about three months.

Fig. 2 shows two common standard firebricks made of Kentucky clay, and one in the center which was not. The Kentucky clay bricks were of the specification calling for a stiff mud, de-aired, hard pressed, dense surface brick. The brick are shown just as they were taken from the furnace, aside from some mechanical damage caused in removing them. The two Kentucky clay bricks show practically no signs of disintegration.

Fig. 3 shows another view of the same three bricks. The center brick is definitely not acceptable. It is a dry press brick that had not been given a reducing burn. The other two, particularly the one on the top, showed a small amount of deposition on the surface, but did not contain any large formations of carbon whatsoever.

Fig. 4 shows a second quality clay brick. All of these bricks became badly damaged in the tests and in some cases completely powdered away. As a matter of fact, the samples shown here were about the best pieces that could be obtained. The carbon depositions were large and numerous. They appeared on all surfaces of the broken pieces.

Fig. 5 shows three of the cheapest firebricks made, glazed on five sides and placed in the furnace resting on the unglazed side. This view also shows one light weight refractory brick, glazed. In every test the glazed brick, in spite of quality, stood up quite satisfactorily. The unglazed surface was slightly purple, but that was the only sign of carbon deposition.

Fig. 6 shows two grades of light weight refractory bricks, not

glazed. These bricks stood up just as well without a protective coating as with one. There were some carbon deposits throughout the brick, but they were small and well distributed. The high porosity of this light weight brick seems to protect it from disintegration.

In the final selection, after eliminating the light refractory on account of abrasion troubles, it appeared that either glazed standard firebrick, or the stiff mud Kentucky clay firebrick should be used. Naturally, the additional cost for glazing threw out all firebricks that required such a treatment. The only one left was the Kentucky stiff mud, dense surface brick.

Several types were found passable. For our purpose, however, only one hard burned, Kentucky clay brick proved wholly satisfactory because it showed no signs of disintegration whatsoever.

Louis J. Trostel,⁵ before the Institute of American Chemical Engineers, states:

"Successful de-airing of stiff mud brick by vacuum has extended rapidly since 1932 in the industry and the number of installations is more general than for the de-airing of dry-press brick. In the stiff mud de-airing process the shredded clay is fed from a plug mill into a chamber maintained under a vacuum of about 25 inches of mercury then extruded as a continuous column by an auger through the die which forms the ware. Improved density and nearly complete elimination of the lamination planes typical of products made by this molding process are accompanied by a very high increase in strength."

With brick manufactured in this manner, it is evident that they will be denser, of lower porosity, and, consequently, more resistant to gas penetration. Another redeeming virtue is that they are both stronger and heavier. And this means that their heat storage capacities will be greater.

All bricks of this particular brand are given a reducing firing period at the end of their fire. This causes the iron oxide segregations to change into the magnetic type, and then compels this oxide and metallic iron to combine with the silica in the brick to form an inert glass.

These experiments were merely repetitions of what the blast furnace operators had been forced to do years ago, but to the average heat treating furnace user and builder they were as new as they were enlightening.

⁵Louis J. Trostel, "New Developments and Trends in Refractories Processes and Materials," Wilmington, Del., Meeting, May 13-15, 1935. *Transactions, American Institute of Chemical Engineers*, Vol. 31, No. 1, Dated March 25, 1935, p. 477.

CONCLUSIONS

It would seem, then, that when the furnace user or designer is faced with a problem such as this one involving the disintegration of firebrick in a protective gas atmosphere, it would be well for him to consider if there is, or ever has been, an analogy of his problem in some other field. It might be well for him to take the materials supplier more completely into his confidence, weighing with the utmost care all suggestions and recommendations.

To date, very little thought has been given to the choice between firebrick and light refractories in high carbon steel protective atmosphere furnaces. The light weight materials have seldom been challenged in this particular field. Firebrick manufacturers, however, tell us that they are now becoming awakened to this new market for their product and that they intend to acquaint the industry with all comparative facts. It is not a large market that we speak of, but it is a growing one and its cultivation in the protective atmosphere furnace field will work more to the advantage of the brick and furnace user than to the brick supplier.

FURTHER NOTES ON PRECIPITATION HARDENING IN THE HEAVY ALLOYS

By W. P. SYKES

Abstract

Reporting the continuation of previous investigations this paper is concerned with further aspects of the precipitation hardening phenomena exhibited by binary iron-tungsten and iron-molybdenum alloys. The subject matter includes:

(1) Hardness changes at elevated temperatures accompanying precipitation and attainment of structural equilibrium at these temperatures.

(2) Comparison of single phase with two phase alloys in respect to behavior during precipitation hardening at intermediate temperatures.

(3) Effect of cooling rate from solution treatment upon precipitation hardening characteristics.

INTRODUCTION

A STUDY of some of the factors which influence the precipitation hardening performance of single phase binary iron-tungsten and iron-molybdenum alloys has been described in a recent paper (5).¹ To determine the possible effects of certain other factors this investigation has been continued with the results reported in the following pages.

The subject matter is grouped under three headings as follows:

I—Precipitation at Elevated Temperatures—The temperatures employed in this portion of the work were such that by the process of precipitation the alloys approached a state of structural equilibrium within a period of 100 hours.

II—Precipitation Hardening in Two-Phase Alloys at Intermediate Temperatures—This is essentially a comparison of a single phase solid solution of fixed concentration with a two-phase alloy, the solid solution phases being approximately of equal concentration.

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented before the Twenty-second Annual Convention of the Society held in Cleveland, October 21 to 25, 1940. The author, W. P. Sykes, is metallurgist, Cleveland Wire Works, General Electric Co., Cleveland. Manuscript received June 18, 1940.

III—Effect of Cooling Rate from the Solution Treatment—

Single phase solid solutions as water quenched are compared with the same as cooled in air in respect to their performances during subsequent precipitation hardening.

PRECIPITATION AT ELEVATED TEMPERATURES

Among the factors contributing to precipitation hardening performance should be considered (a) hardness of the precipitated phase (b) particle size of the precipitated phase and (c) volume of the precipitated phase.

It might be expected that such effects would be most noticeable during the final stages of the aging cycle when the hardness is well past a maximum and the solid solution is approaching a concentration

Table I
Description of Alloys

Mark	Per Cent Tungsten		Per Cent Molybdenum		
	By Weight	By Atom	By Weight	By Atom	
5 W	16.0	5.5	Si—Max. 0.12 Per Cent
7 W	20.2	7.2	Mn—Max. 0.20 Per Cent
9 W	23.9	8.8	C—Max. 0.02 Per Cent
5 Mo	8.9	5.4	
9 Mo	14.1	9.8	

Alloys melted under a basic slag in the magnesia crucible of an induction furnace, using Armco Ingot Iron, high purity tungsten and molybdenum.

The ingots, 1.5 inches in diameter, were hot swaged to a diameter of 0.5 inch.

which is in equilibrium with the precipitated phase at the temperature of the precipitation treatment. During this period both precipitation and agglomeration or particle growth are in progress, both being functions of the diffusion rate of the solute in the solvent metal.

The temperatures commonly employed in precipitation hardening are so low that the solid solution would be expected to reach equilibrium concentration only after extended periods of treatment. This fact renders difficult the evaluation of the effects contributed to the properties of the conglomerate by (a) the solid solution and (b) the precipitated phase. At elevated temperatures, however, structural equilibrium may be approximated more speedily and from the resulting structures it should be possible to obtain some indication of the hardening effect of the second phase.

Such a comparison of three compositions in the binary iron-

tungsten system is shown in Fig. 1. The curve ABC will be recognized as the solvus line at the iron-rich end of the system, as previously determined (6) and the Rockwell D hardness of each of the three compositions as water quenched from the alpha field is marked on the chart at A, B and C. The preparation of these alloys has been described in detail elsewhere (5) and is briefly reviewed in Table I.

In order to compare the hardness of a solid solution of known composition with a conglomerate made up of the same solid solution and fixed quantity of a second phase the following treatment was employed.

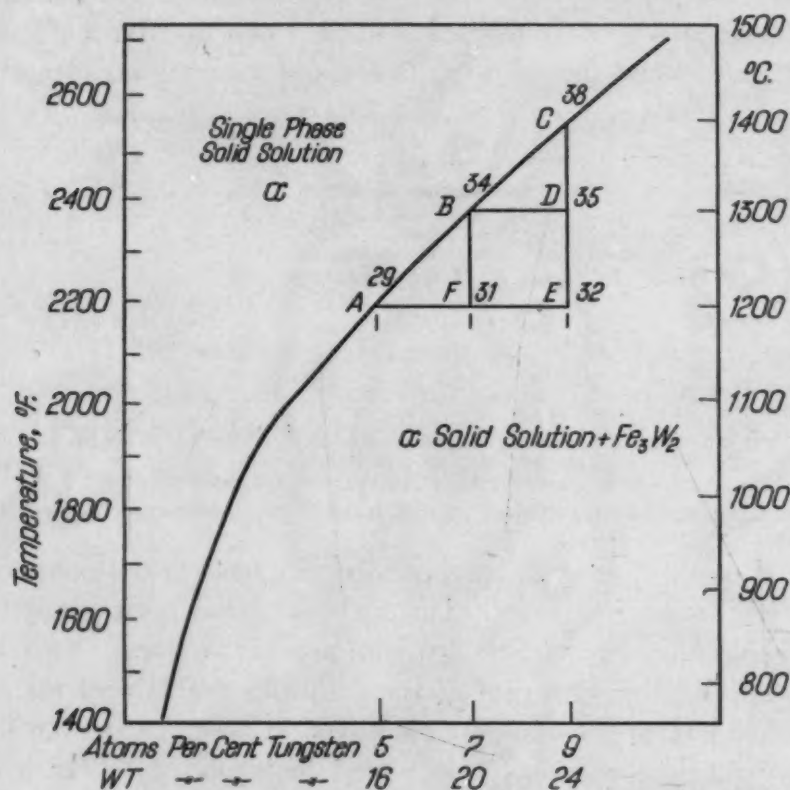


Fig. 1—Comparison of Hardness (Rockwell D) in Single Phase and Two-Phase Structures. Points A, B and C Represent Single Phase Solid Solutions of 5, 7 and 9 Atomic Per Cent Tungsten Respectively. Points D, E and F Represent Two-Phase Structures Approximating Equilibrium at Their Respective Temperatures.

A set of specimens of 9 W (24 per cent tungsten by weight) was water quenched from 1450 degrees Cent. (2640 degrees Fahr.) a temperature well above the solvus for this composition. These were then heated in the hydrogen atmosphere of a tungsten wound resistor

furnace at 1200 ± 15 degrees Cent. (2190 degrees Fahr.) for periods ranging from 2 minutes to 100 hours and quenched. The changes in hardness and in microstructure, indicate that structural equilibrium has been approximated after some 50 hours at 1200 degrees Cent. (2190 degrees Fahr.). In Fig. 2 the curve marked 9 W—1200 indicates the hardness change throughout the heating period.

In order to insure comparable heating rates the individual specimens ($\frac{1}{2}$ inch in diameter and $\frac{3}{8}$ inch long) were placed in the furnace separately for the shorter intervals (2 minutes to 30 minutes). Obviously the maximum hardness developed during the first few minutes reflects the precipitation which occurred before the alloy had attained the temperature of the furnace, as indicated on the chart.

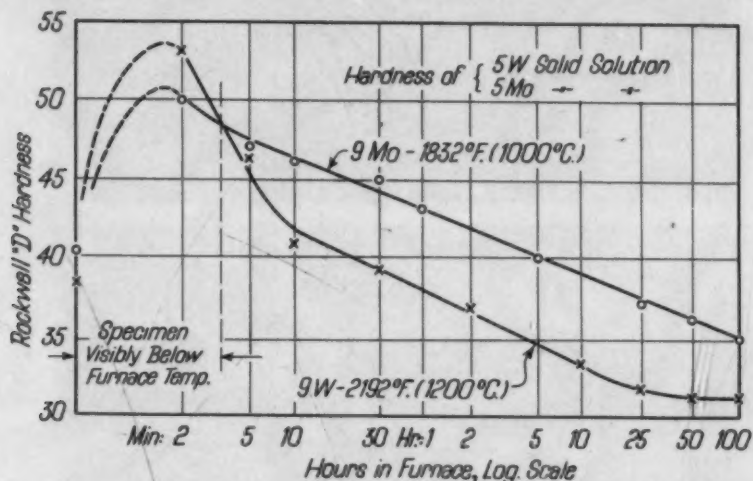


Fig. 2—Hardness-Time Relationships in Binary Iron-Tungsten and Iron-Molybdenum Alloys Approaching Structural Equilibrium at the Temperatures Indicated on the Curves.

At the end of fifty hours at 1200 degrees Cent. (2190 degrees Fahr.) the structure of alloy 9 W consists of a solid solution of approximately 16 per cent tungsten by weight together with the excess 8 per cent tungsten carried in the relatively massive crystals of Fe_3W_2 as shown in Fig. 7. The hardness of alloy 9 W during this period has decreased from $R_D 38$ to 32 and exceeds by only 3 points that of the solid solution 5 W.

Crystal growth of the precipitated phase concurrent with softening may be observed in the series of photomicrographs shown in Figs. 3 to 7 inclusive, the first two having the higher magnification of 500 diameters. The rate of particle growth is obviously extremely high during the first 10 minutes while after 25 hours no growth was detectable from casual inspection.

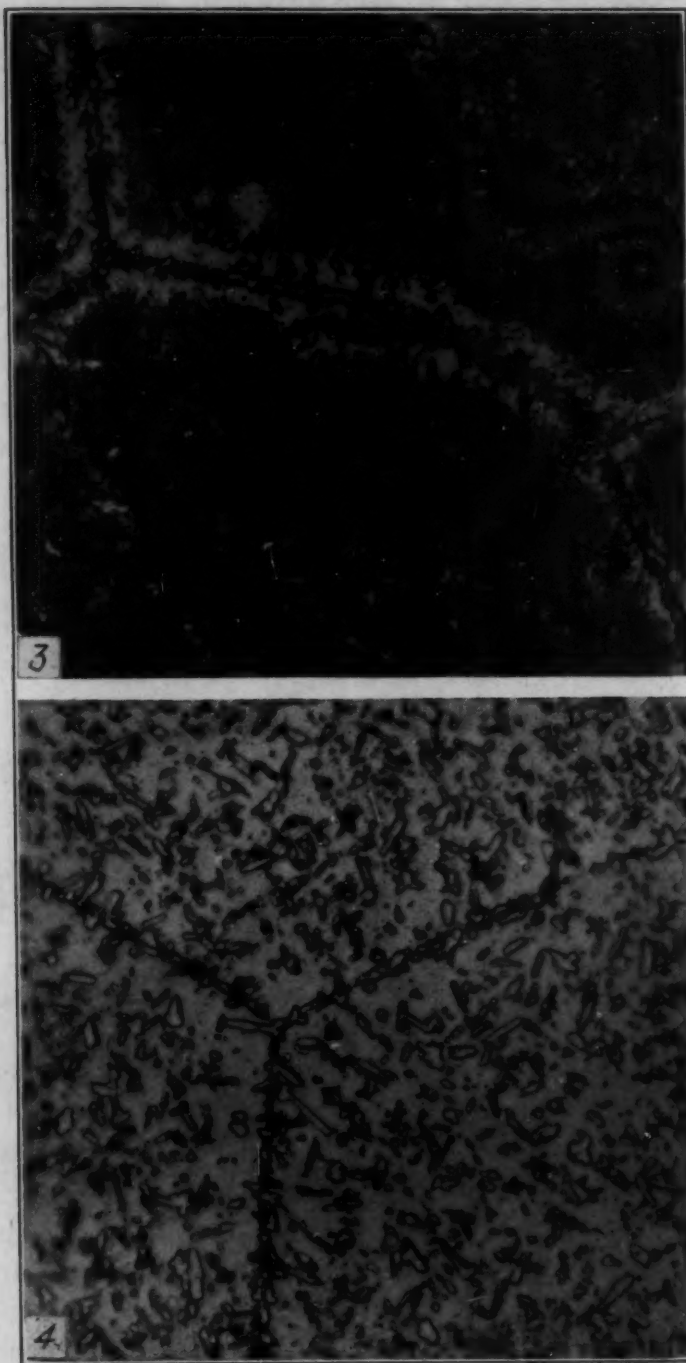


Fig. 3—Alloy 9 W. Iron + 24 Per Cent Tungsten by Weight. Water-Quenched from 1450 Degrees Cent. and Heated for 5 Minutes in Furnace at 1200 Degrees Cent. Water-Quenched from 1200 Degrees Cent. D 46. $\times 500$. The Etchant Used in all Cases Consisted of Four Parts H_2O_2 + One Part H_3PO_4 .

Fig. 4—Alloy 9 W. Water-Quenched from 1450 Degrees Cent. and Heated for 10 Minutes at 1200 Degrees Cent. D 41. $\times 500$.

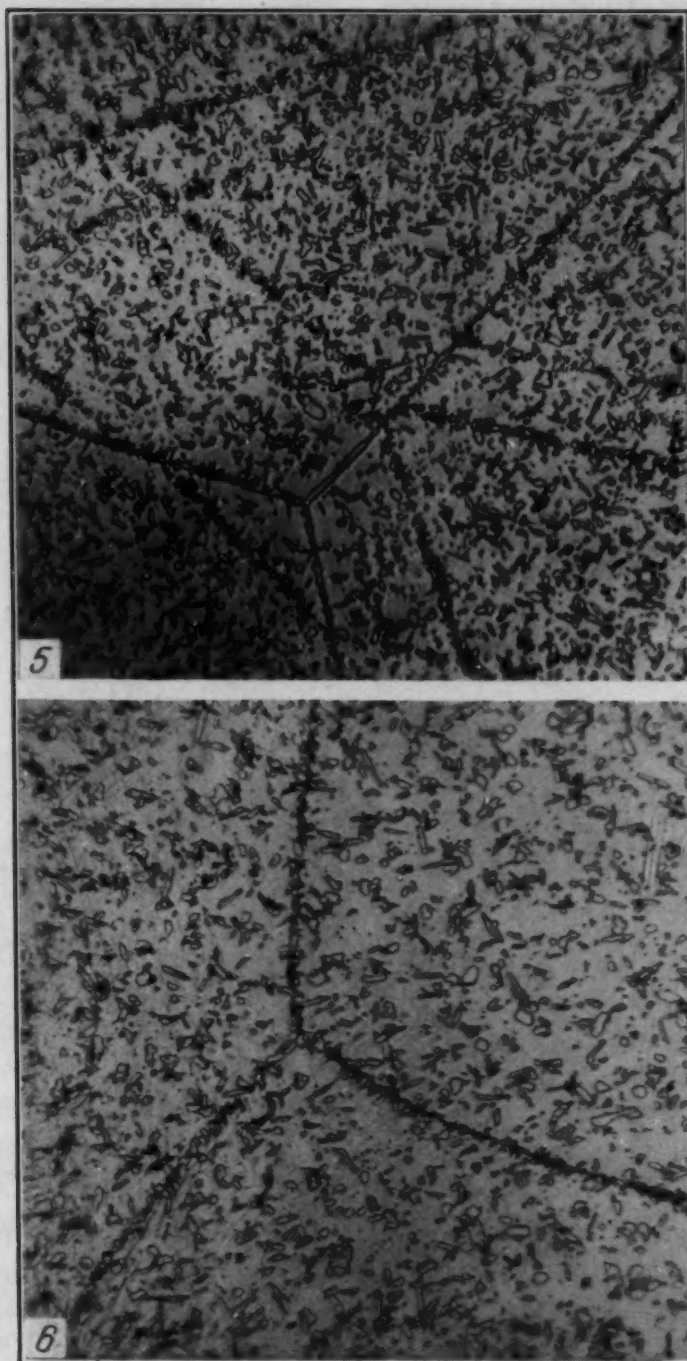


Fig. 5—Alloy 9 W. Water-Quenched from 1450 Degrees Cent. and Heated for 30 Minutes at 1200 Degrees Cent. D 39. $\times 200$.

Fig. 6—Alloy 9 W. Water-Quenched from 1450 Degrees Cent. and Heated for 2 Hours at 1200 Degrees Cent. D 36. $\times 200$.

Results of two analogous comparisons involving 5 W, 7 W and 9 W are included in Fig. 1 and illustrate the minor hardening effect of relatively large crystals Fe_3W_2 in the presence of the alpha solid solution.

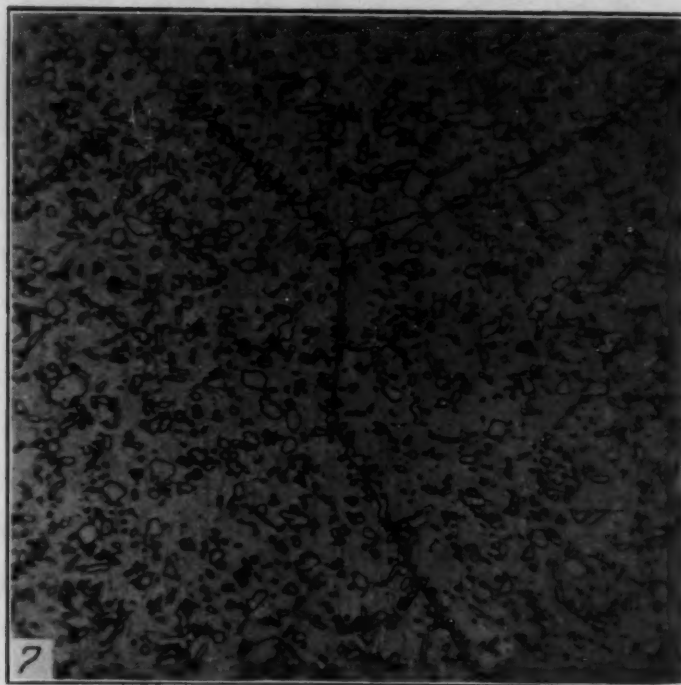


Fig. 7—Alloy 9 W. Water-Quenched from 1450 Degrees Cent. and Heated for 50 Hours at 1200 Degrees Cent. D 32. $\times 200$.

In the iron-molybdenum system the solid solution saturated with 5 atomic per cent molybdenum (8 per cent by weight) has a solvus temperature of about 1000 degrees Cent. (1830 degrees Fahr.). The 9 atomic per cent alloy in this system after water quenching from $1250 \pm$ degrees Cent. (2280 degrees Fahr.) a temperature within the alpha field, was heated at 1000 ± 20 degrees Cent. (1830 degrees Fahr.) for periods as long as 100 hours. As shown by curve 9 Mo—1000 in Fig. 2 this alloy had not attained a state of structural equilibrium at the end of this period, showing a hardness of 35 as compared to 29, characteristic of the 5 Mo solid solution.

Even though the diffusion rate of molybdenum is higher than that of tungsten in alpha iron the lower temperature of precipitation to which the molybdenum alloys were subjected appears to offset this fundamental difference in diffusion rates.

PRECIPITATION HARDENING IN TWO-PHASE ALLOYS

From studies of resistance changes in two-phase aluminum-copper alloys Hartnagel concluded (1) that the presence of a second phase substantially increased the strains resulting from quenching or cooling to sub zero temperatures.

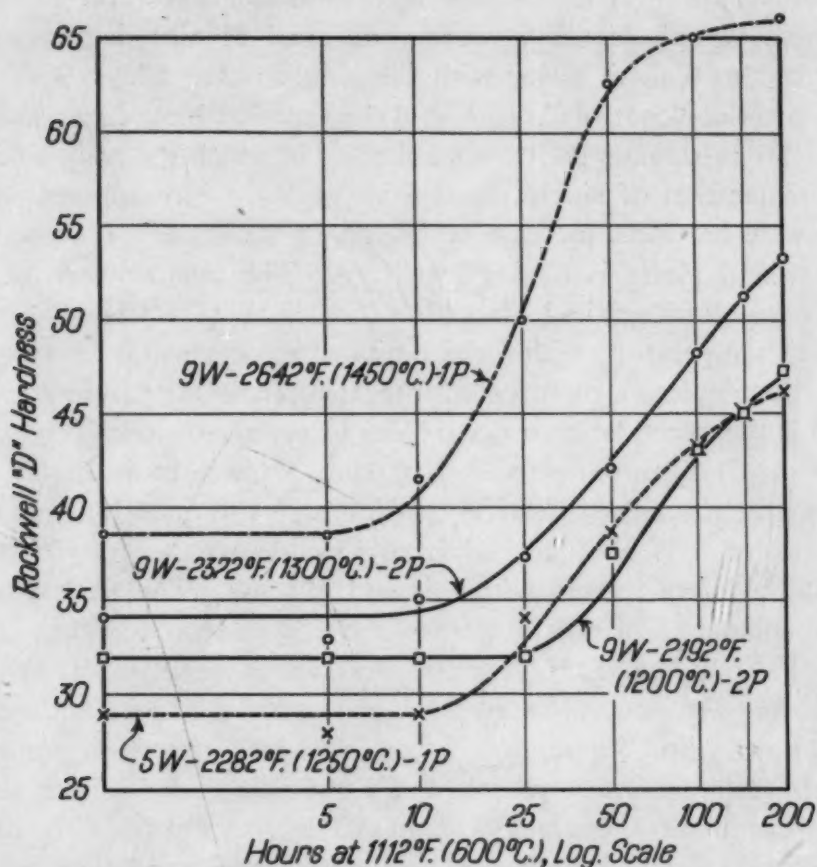


Fig. 8—Time-Hardness Curves for Single Phase and Two-Phase Alloys of Iron and Tungsten. 9 W—1450—1 P. Single Phase Alloy Water-Quenched from 1450. 9 W—1300—2 P. Two-Phase Alloy in Which Solid Solution Has Tungsten Concentration of 7 Atomic Per Cent (20 Per Cent by Weight). 9 W—1200—2 P. Two-Phase Alloy in Which Solid Solution Has Tungsten Concentration of 5 Atomic Per Cent (16 Per Cent by Weight). 5 W—1250—1 P. Single Phase Alloy Having Concentration of 5 Atomic Per Cent, all Water-Quenched from the Indicated Temperatures Before Aging at 600 Degrees Cent.

It appeared possible that a similar effect if present in the two-phase iron-tungsten alloys might result in a more rapid hardening in the two-phase alloy during a precipitation treatment subsequent to the quenching operation.

In order to test this possibility three series of 9 W specimens

were treated in the following manner. After water quenching from 1450 degrees Cent. (2640 degrees Fahr.) one group was heated for 50 hours at 1300 degrees Cent. (2370 degrees Fahr.) and water quenched to produce a two-phase structure in which the tungsten concentration of the solid solution approximated that in single phase 7 W. A second group was heated for 50 hours at 1200 degrees Cent. (2190 degrees Fahr.) to compare with solid solution 5 W.

These two-phase alloys were then aged at 600 degrees Cent. (1110 degrees Fahr.) along with the single phase alloys 9 W and 5 W to serve as "controls". In Fig. 8 the time-hardness curve marked 1200 (2 P) represents the two-phase alloy in which the solid solution approximates that of single phase 5 W (1 P). Any anticipated acceleration in hardness increase which might accompany the presence of the second phase is obviously absent. The performance of the single phase alloys marked 1450 (1 P) and 5 W (1 P) is normal for this aging temperature within the limits of experimental error.

In the duplex structures of the two-phase alloys under consideration the second phase occurred as crystals of relatively large dimension. It appeared possible that such a phase more highly dispersed might act more effectively to increase the quenching strains. Two series of 9 W as quenched from 1450 degrees Cent. (2640 degrees Fahr.) were treated with this end in view. The first, after a water quench, was placed in a furnace at 1200 degrees Cent. (2190 degrees Fahr.), held for 2 minutes and water quenched from this temperature. The second series was aged at 650 degrees Cent. (1200 degrees Fahr.) for 30 minutes following which treatment some of the specimens were water quenched and some air cooled. Both series were then aged at 600 degrees Cent. (1110 degrees Fahr.) with results which offered no indication of accelerated age hardening in the quenched specimens or measureable effect in ultimate hardness due to the difference in cooling rates.

EFFECT OF COOLING RATE FROM SOLUTION TREATMENT ON PRECIPITATION HARDENING PERFORMANCE IN SINGLE PHASE ALLOYS

A marked effect on the rate and magnitude of precipitation hardening in aluminum-copper alloys has recently been reported by Hunsicker (2). His data show that high purity aluminum-copper alloys and aluminum-copper-iron alloys quenched in boiling water

age at a higher rate and to a greater extent at room temperature than when quenched in cold water.

Preliminary investigations of a like nature concerning such effects in iron-tungsten and iron-molybdenum alloys have yielded data which is embodied in Figs. 9 and 10. All the specimens were of the dimensions previously stated. Those marked W. Q. in Figs. 9 and 10 were quenched from the solution treatment in tap water at

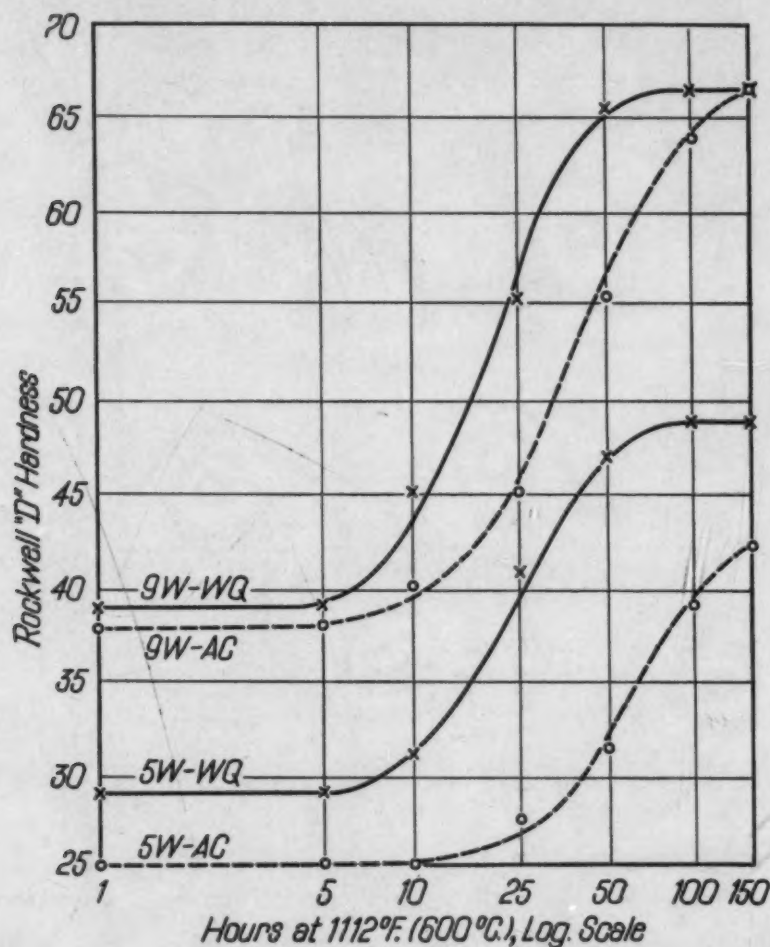


Fig. 9—Time-Hardness Curves for Iron-Tungsten Alloy of Two Compositions, as Water-Quenched and as Air-Cooled from Temperatures of Solution Treatments. (1450 and 1250 Degrees Cent.).

about 25 degrees Cent. The specimens marked A. C. were placed on the furnace top and cooled at such a rate as to lose color in about 45 seconds.

In all cases the water quenched alloys were harder than those which were cooled in air. In all cases, moreover, at the aging tem-

peratures employed the alloys in the water quenched condition hardened more rapidly than those which had been cooled in air. In the alloys of the higher solute concentration, 9 W and 9 Mo, both the water quenched and the air cooled groups develop an equal hardness

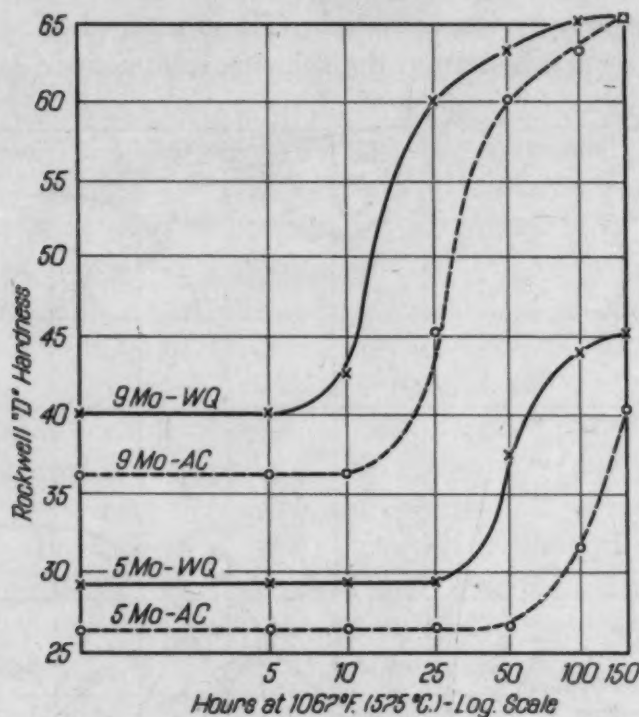


Fig. 10—Time-Hardness Curves for Iron-Molybdenum Alloys of Two Compositions, as Water-Quenched and as Air-Cooled from Temperature of Solution Treatment. (1250 Degrees Cent.).

within an aging period of 100 hours in contrast to the alloys of lower concentration.

The data presented in Figs. 8, 9 and 10 were obtained from duplicate specimens of each alloy, aged, in each case, at the same time and placed in the furnace within a distance of 1 inch from the thermocouple.

The performance of the iron-tungsten and iron-molybdenum alloys during the early stages of precipitation hardening at 575 to 600 degrees Cent. (1065 to 1110 degrees Fahr.) are markedly influenced by a difference of 10 degrees Cent. in the average furnace temperatures. In order, therefore, to obtain a significant comparison of the hardening rates throughout this period certain precautions must be observed. For example, the specimens to be compared at one tem-

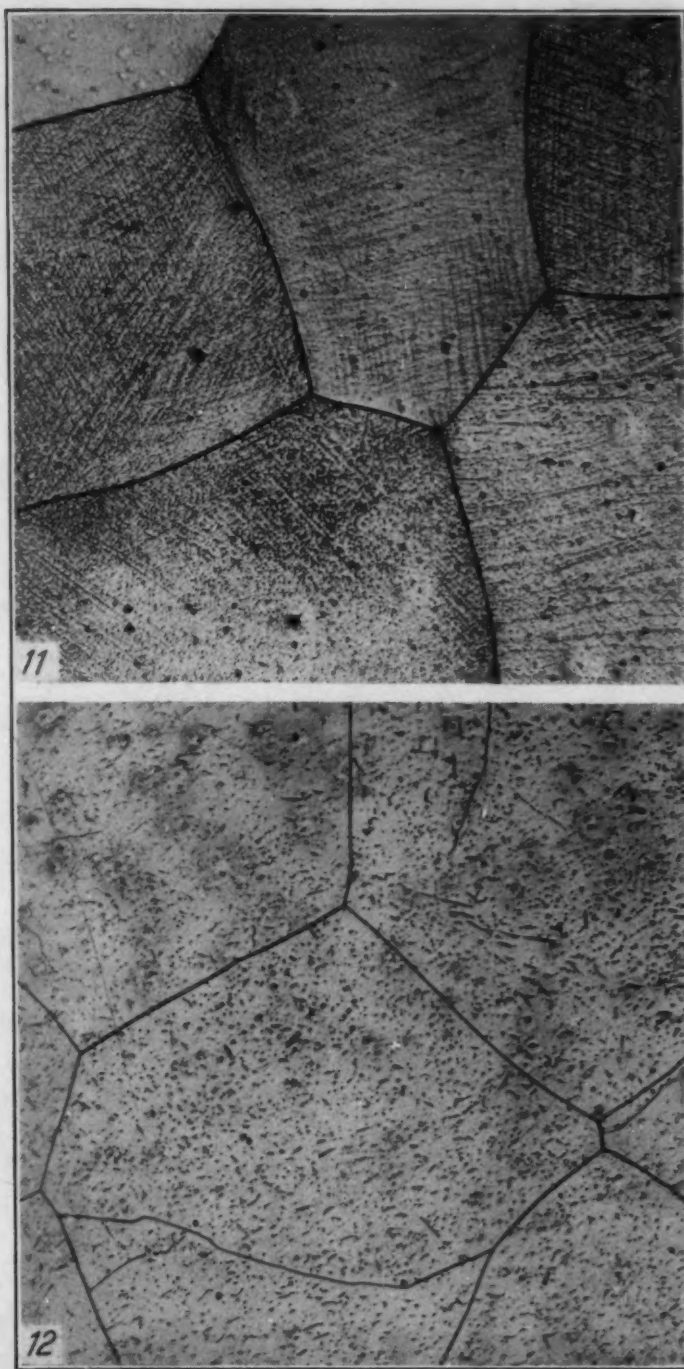


Fig. 11—Alloy 7 W. Iron + 20 Per Cent Tungsten by Weight. Water-Quenched from 1350 Degrees Cent. and Aged at 600 Degrees Cent. for 5 Hours. D 35. $\times 100$.

Fig. 12—Alloy 7 W. Air-Cooled from 1350 Degrees Cent. and Aged at 600 Degrees Cent. for 5 Hours. D 28. $\times 100$.

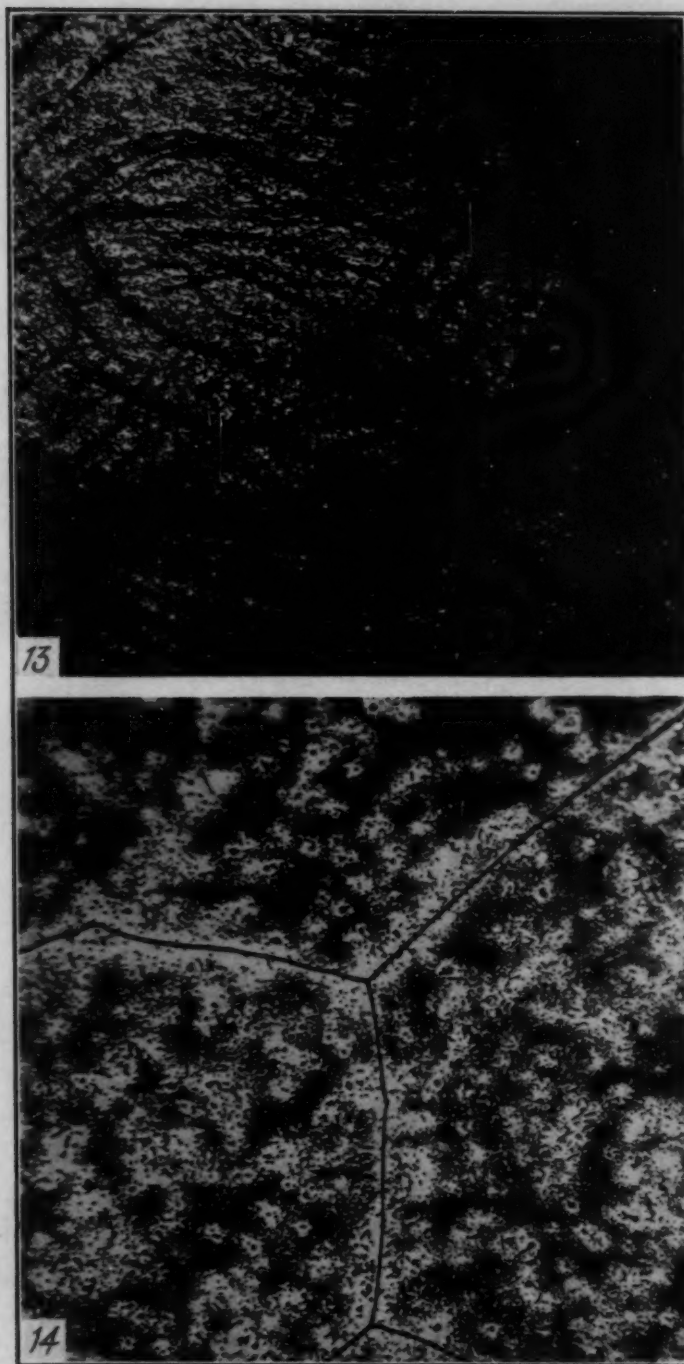


Fig. 13—Alloy 7 W. Water-Quenched from 1350 Degrees Cent. and Aged for 50 Hours at 600 Degrees Cent. D 50. \times 100.

Fig. 14—Alloy 7 W. Air-Cooled from 1350 Degrees Cent. and Aged for 50 Hours at 600 Degrees Cent. D 39. \times 100.

perature should be aged simultaneously in a single furnace. They should, moreover, be grouped about and as near as possible to the hot junction of the thermocouple. Many of the discrepancies between initial hardening rates shown by an identical alloy when aged at different times at the same nominal temperature may undoubtedly be traced to a slight difference in the actual temperatures.

The foregoing data while obviously incomplete suggest that variation in cooling rate subsequent to solution treatment is an important factor in determining the rate of hardening at least during the early stages of precipitation at the lower aging temperatures for a given alloy.

That the cooling rate determines the mode of subsequent precipitation is strikingly demonstrated by the microstructures in Figs. 11, 12, 13 and 14. The linear pattern which brands the water quenched alloy as previously reported (5) is apparent in Figs. 11 and 13. In marked contrast appears the nuclear form of precipitate in the same alloy aged for 5 hours and for 50 hours after air cooling from the solution treatment as shown in Figs. 12 and 14. These structures support Hunsicker's postulate regarding the "nucleation" or "seeding" which occurs during the slower quench.

The relative hardening rates upon subsequent aging are, however, quite the reverse of those found by Hunsicker in that the more rapidly cooled specimens harden at a higher rate. Kempf as well as Fink and Smith have observed in aluminum alloys the phenomena reported by Hunsicker.

Obviously the experimental data pertaining to this aspect of precipitation hardening are far too meager to serve as a basis for any generalizations at the present time.

SUMMARY

The following observations would appear to briefly summarize the foregoing data:

1. The precipitation hardening characteristics of a two-phase alloy seem to be identical with those of the solid solution phase.
2. No evidence is apparent to indicate that the presence of a second phase either highly dispersed or in massive form introduces sufficient additional strain during the quench to alter the rate of precipitation hardening.

3. The solid solutions as water quenched from the solution treatment are measurably harder than after cooling in air.

4. The solid solutions as water quenched harden more rapidly during subsequent aging than do the same alloys in the air cooled condition.

Bibliography

1. W. Hartnagel, "Die Kalt-Aushärtung von Aluminum-Kupfer Legierungen," *Zeit. für Metallkunde*, Vol. 30, 1938, No. 3, p. 81.
2. H. Y. Hunsicker, "Precipitation Hardening Characteristics of High Purity Aluminum-Copper and Aluminum-Copper-Iron Alloys." SYMPOSIUM ON AGE HARDENING OF METALS, American Society for Metals, 1939, p. 56.
3. W. L. Fink and D. W. Smith, "Age-hardening of Aluminum Alloys, III—Double Aging Peaks," *Transactions*, American Institute of Mining & Metallurgical Engineers, Vol. 128, 1938, p. 223.
4. L. W. Kempf, discussion of paper by A. Phillips and R. M. Brick, *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. III, 1934, p. 92-112.
5. W. P. Sykes, "Precipitation Hardening in the Heavy Alloys." SYMPOSIUM ON AGE HARDENING OF METALS, American Society for Metals, 1939, p. 82.
6. W. P. Sykes, "Notes on the Solidus Temperatures in the Systems Iron-Tungsten and Iron-Molybdenum," *TRANSACTIONS*, American Society for Metals, Vol. 24, 1936, p. 541.

DISCUSSION

Written Discussion: By H. Y. Hunsicker, research laboratories, Aluminum Company of America, Cleveland.

As Dr. Sykes has pointed out, there is a dearth of experimental data adaptable to a thorough analysis of the quenching rate factor in precipitation hardening. The state of the as-quenched alloy as well as the course of subsequent aging events in precipitation hardenable systems must be dependent to a considerable degree upon the rate of cooling through the temperature range not greatly below the solvus wherein the balance between degree of supersaturation and diffusivity result in maximum rate of nucleation.¹ In this same general temperature range the greater plasticity of the alloy creates a susceptibility to quenching strain. Thus, relatively fast rates of quenching through this temperature range promote thermal strain, which may or may not be accompanied by localized nucleation (at slip planes, grain boundaries, etc.), whereas relatively slow quenching rates are conducive to concomitant random nucleation. In fact, the existence of an intermediate quenching rate which results in complete suppression of nucleation as well as freedom from strain is problematical and may vary from system to system.

The expected increase in initial aging rate brought about by quenching strain or localized nucleation resulting from such strain has been well demonstrated. There are also evidences that the hardening may be accelerated by quenching rates which are just slow enough to permit a restricted amount of

¹A temperature range corresponding to the region of fastest reaction or "nose" of the isothermal diagram for austenite decomposition.

finely dispersed, random nucleation (Reference 2). As the quenching rate becomes slower than this, the nuclei form at increasingly higher temperatures, are therefore larger and decreasingly efficacious in promoting hardening since they may pass through the "critical" size at a very early period in the aging cycle. Moreover, early precipitation of sufficient magnitude may reduce the degree of supersaturation of the solid solution to effect a decrease in the rate of nucleation and growth at the aging temperature. The net result is, therefore, a lower hardening rate.

The microstructures of the air-cooled and aged iron-20 per cent tungsten alloy (Figs. 12 and 14) reveal large precipitate particles which apparently grew rapidly from large (high temperature) quenching nuclei and because of their size failed to contribute to the hardening. Furthermore, at the state represented by Fig. 14 the light areas surrounding the large precipitate particles as well as adjacent to the grain boundaries apparently have been greatly impoverished in solute atoms and probably are not contributing to the increase in hardness resulting from normal precipitation. This evidence indicates that in the case of these alloys an air cool is too slow to afford a degree of high temperature nucleation favorable to increased hardening rate. Indeed, it would be quite fortuitous if a randomly selected quenching medium should provide the necessary cooling rate, although for aluminum-copper alloys a boiling water quench probably approaches the optimum quenching rate rather closely. It is suggested that quenching rates intermediate to those reported for the iron-tungsten and iron-molybdenum alloys may bring about increased rate and magnitude of precipitation hardening through a refinement in the size and dispersion of the quenching nuclei.

The effect of formation of nuclei during the quench on the ultimate mechanical properties of a precipitation hardening alloy is not certain. Although it may be reasoned that either localized (strain) or random nucleation or thermal strain unaccompanied by formation of nuclei should result in a wider distribution of particle sizes, and, therefore, fewer particles simultaneously reach the "critical" size, the existing experimental data fail to reveal a consistent impairment in the maximum properties reached during the aging cycle.³

At Dr. Sykes' request microhardness measurements were made on an iron-30 per cent tungsten alloy which had been brought to approximate structural equilibrium at 1200 to 1225 degrees Cent., through the following treatment: held 30 minutes at 1500 degrees Cent., cooled from 1500 to 1225 degrees Cent. over a 75-hour period, held at 1200 to 1225 degrees Cent. for 25 hours, and water quenched. The structure consisted of only two phases; the solid solution matrix containing about 16 per cent tungsten and massive crystals of Fe_3W_2 . The approximate microhardness values (Bierbaum) for the two phases are as follows:

	λ Scratch Width, Microns	K Microhardness
Solid Solution Matrix	6.5-7.4 μ	185-240
Fe_3W_2	1.4-1.5 μ	4400-5100

These values may be compared with those for Armco Iron ($K = 240$) and cementite ($K = 5600$) for a ready comprehension of the relative hardness.

³M. U. Cohen, correspondence, Reference 3. Note also Figs. 9 and 10, this paper.

Author's Reply

I wish to thank Mr. Hunsicker for his kind collaboration. He has ably explained away the discrepancy which might appear to exist between the effects of cooling rates in the cases of aluminum-base alloys and the alloys of iron-tungsten and iron-molybdenum. His application of the microcharacter method in comparing the hardness of the two phases present in these duplex structures illustrates nicely the unique function of this instrument. In this case as in many similar instances, it is quite impossible to obtain the intermetallic phase in a form suitable for hardness measurements by other methods. Mr. Hunsicker's contributions add greatly to the value of this paper.

SOME PROPERTIES OF HOT-PRESSED AND SINTERED COPPER POWDER COMPACTS

BY CLAUD G. GOETZEL

Abstract

This study includes the determination of the effect of variations of the following factors:—compacting pressure and temperature, and annealing. The tests on the compacts so produced include density, rolling ability, hardness and compressive strength as well as microscopic examination of the structure of the compacts.

For compressing the compacts, a small hydraulic press of conventional design was used for loading, and special equipment for heating and molding was built. Testing was done according to standard procedures as far as possible.

From these tests the following conclusion has been drawn: High compacting pressures give the highest density and the best physical properties.

The compacts molded at high temperature are found to have a very high density, and to be harder and stronger (compression test) than sintered or cast copper even though the time at high temperature is very short and there is no grain growth and very little recrystallization as shown by the photomicrographs. This definitely disproves the older belief that it is necessary to achieve a crystalline structure resembling that of ordinary copper before satisfactory strength, ductility, and malleability can be secured. This points the way for entirely new developments in the field of powder metallurgy.

THE technique of forming solid objects from metal powders is commonly known as "powder metallurgy". Consolidation is brought about by pressing the powder in a steel mold under heavy load, and then heating the resulting compact to a temperature below the melting point of the material; or by compressing the powder in a heated mold so that both of the above operations are carried out simultaneously. In some cases, the solids so produced are retreated in a manner identical with the first treatment in order to obtain better physical properties. The compacts must be protected from

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oxidation throughout all high temperature operations by maintaining them in a neutral or reducing atmosphere.

Despite the commercial importance of these processes, little systematic research has been done in this field. There is some reliable information available but it is largely confined to special problems, and those investigations of general interest that have been conducted are, in general, rather incomplete. It is therefore the aim of this work to establish a systematic relationship between some variables of powder metallurgical processes and the structure and properties of the resulting compacts.

Copper was chosen for this investigation as the need for more information on the behavior of metals of low and medium melting point was urgent, and because copper is the most prominent of the nonferrous metals.

Although much literature has been published on the subject of powder metallurgy, most of the publications deal with refractory metals or their carbides, and only little work has been done on hot compression of copper powders.

Sauerwald (1)¹ was the first to conduct researches on hot- and cold-compression of metal powders such as iron and copper. He found a definite superiority in tensile strength of hot-compressed specimens, as compared with compacts which were molded at room temperature and subsequently heat treated. Jones (2) summarized the advantages of hot pressing, combined with a suitable atmosphere, as follows:

1. Reduction of gas content
2. Accuracy in size (no shrinkage effects)
3. Higher strength and hardness
4. Higher densities (reduction of porosity)
5. Probably greater elongation
6. Complete sintering at high temperatures within very short heating times.

In practice, however, hot pressing is limited by difficulties in molding at temperatures above a certain limit.

Trzebiatowski (3) investigated the density of hot-compressed copper in an important research on hot-molded copper powder compacts. Chemically pure copper powder of finest grade was compressed in a hydrogen atmosphere at temperature between 20 and

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

600 degrees Cent. (70 and 1110 degrees Fahr.) at 95 tons per square inch. The density rapidly increased up to 200 degrees Cent. (390 degrees Fahr.) where it reached the value of 8.7 grams per cubic centimeter. Between 200 and 400 degrees Cent. (390 and 750 degrees Fahr.) the value rose slowly to a maximum of 8.9 grams per cubic centimeter very closely approaching the density of copper, 8.93 grams per cubic centimeter. Trzebiatowski also studied the electrical resistance of hot-compressed copper compacts. He found that the specific resistance, as well as the temperature coefficient of resistance, were close to the corresponding values for gross copper.

The hardness of hot-compressed compacts was examined by the same investigator. The powder used was very fine, the particles averaging 1 micron in diameter. The specimens were compressed at 95 tons per square inch, while the temperature was varied from 20 to 600 degrees Cent. (70 to 1110 degrees Fahr.).

At 100 degrees Cent. the hardness reached a maximum of 190 Brinell; it remained the same to 200 degrees Cent. (390 degrees Fahr.), and then rapidly decreased to 110 Brinell at 600 degrees Cent. (1110 degrees Fahr.). The maximum hardness obtained was much greater than that of ordinary strained copper. The decrease in hardness at higher pressing temperatures was explained by disappearance of the work-hardened structure, as determined by X-ray examination.

The tensile properties of (a) hot-compressed and (b) cold-compressed and sintered compacts have been compared by Sauerwald (1). No detailed description of the equipment for hot pressing is given, but it is mentioned that the specimens were heated in a protective atmosphere. Hot-pressed compacts were produced at temperatures up to 810 degrees Cent. (1490 degrees Fahr.), and at a pressure of 24 tons per square inch. No information on the grade of powder or time of pressing is given. The tensile strength was measured for different temperatures of compacting and the results are summarized in Table I.

Table I
Sauerwald's Tensile Test Results for Sintered and Hot-Compressed Copper

Temperature of Heating	Tensile Strength Cold Pressed and Heated Pounds Per Sq. In.	Tensile Strength Simultaneously Pressed and Heated Pounds Per Sq. In.
610 Degrees Cent. (1130 Degrees Fahr.)	20,400	37,400
715 Degrees Cent. (1320 Degrees Fahr.)	18,750	34,200
810 Degrees Cent. (1490 Degrees Fahr.)	14,650	33,300

RAW MATERIAL

A commercial electrolytic copper powder was used in this investigation as the raw material for making test specimens. Before being compacted to specimens, the powder was reduced at 300 degrees Cent. (570 degrees Fahr.) for 2 hours in a steady stream of commercial hydrogen.

The chemical and screen analyses of the electrolytic copper powder are given in Tables II and III. The photomicrograph Fig. 1 shows the size and shape of the powder particles. For microscopic examination, the particles were mixed with a mounting cement, placed in small cups, and finely polished. For etching the polished sections, ammonium hydroxide plus hydrogen peroxide or ammonium persulphate were used.

Table II
Screen Analysis of Electrolytic Copper Powder Used

Standard Tylor Screen Size		Percentage
Through 20 mesh, on 20 mesh		0
Through 60 mesh, on 60 mesh		0
Through 100 mesh, on 100 mesh		0
Through 150 mesh, on 150 mesh		5.86
Through 200 mesh, on 200 mesh		18.66
Through 250 mesh, on 250 mesh		2.33
Through 325 mesh, on 325 mesh		25.37
Through 325 mesh		47.44

Table III
Chemical Analysis of Impurities in Electrolytic Copper Powder Used

Element	Percentage
Oxygen (Hydrogen Losses)	0.0915
Iron	Trace
Lead	0.023
Antimony	Trace
Tin	Trace
Zinc	Trace
Silica	0.011
Sulphur }	0.057
Carbon }	
Grease	0.029
Nitric Acid Insoluble	0.068

The particles have the nodular or dendritic shape typical for electrodeposited metals. Each particle consists of several individual grains, sometimes spreading from the center to the outer edge; thus, each particle is an aggregate of crystals.

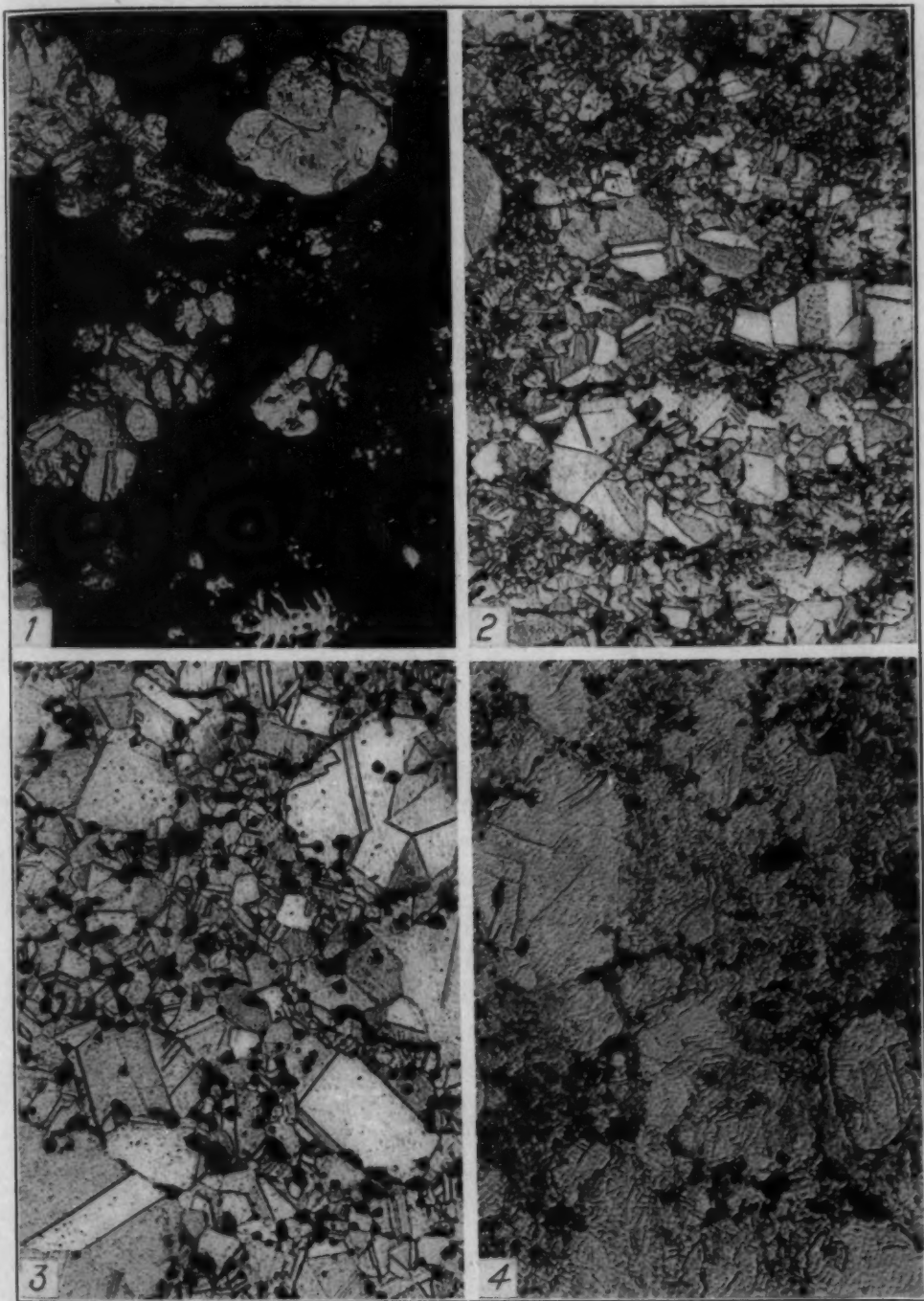


Fig. 1—Electrolytic Copper Powder. $\times 200$.

Fig. 2—Same Powder Cold-Compacted at 50 Tons per Square Inch and Sintered at 600 Degrees Cent. for 1 Hour in Hydrogen. $\times 200$.

Fig. 3—Same Powder Cold-Compacted at 50 Tons per Square Inch and Sintered at 800 Degrees Cent. for 1 Hour in Hydrogen. $\times 200$.

Fig. 4—Same Powder Hot-Compacted at 50 Tons per Square Inch and 300 Degrees Cent. $\times 200$.

PROCESSING METHOD

The apparatus for the hot compression experiments could be operated at temperature up to 500 degrees Cent. (930 degrees Fahr.), and was fitted for compression of the compacts in a reducing atmosphere. Pressure, varying from 5 to 50 tons per square inch was applied by means of a 25-ton hand-operated hydraulic press. Heat was supplied by an electric furnace surrounding the die. The die was built from 18-4-1 tungsten-chromium-vanadium high speed steel which retained its hardness up to 500 degrees Cent. (930 degrees Fahr.). A base-plate, slide and punch were built from the same material as the die. All these parts with the exception of the base-plate were hardened. The base-plate rested on a soft steel adapter, and this, in turn, on a steel collar in a water basin. The base-plate, slide and adapter had eccentric holes of such diameter as to permit dropping the specimens in the quenching bath. The temperature could be maintained constant within 10 degrees Cent. by a rheostat connected in series with the furnace. The temperature was measured by an iron-constantan thermocouple, the junction of which was brought as close to the specimens as possible.

The specimens were pre-compressed cold at 2 tons per square inch into pills of $\frac{9}{16}$ inch in diameter, so that they could easily be dropped into the well of the hot die, $\frac{5}{8}$ inch in diameter. The pills were kept under hydrogen in the die for 5 minutes before applying any pressure in order to insure a uniform temperature. In consideration of the large mass of the die and the small mass and better conductivity of the copper compacts, this time was assumed to be sufficient. After this period, the load was applied, left for 1 minute, and released. The hot compact, upon ejection, was dropped into the water bath.

For subsequent annealing the specimens were placed in an electric muffle furnace under a hydrogen atmosphere. The annealing temperature was kept constant at 800 degrees Cent. (1470 degrees Fahr.) (within 20 degrees Cent.) by means of a rheostat, and the temperature was measured with an iron-constantan thermocouple. The annealing time was 1 hour.

For the sintered specimens, compression of the powder was carried out in hardened steel dies at room temperature. The compacts were molded into a shape suitable for testing which made subsequent machining superfluous. The pressure was varied from 5 to 50 tons

per square inch and was supplied by a 50-ton hand operated hydraulic press.

The sintering operations were carried out in a tubular electric resistance furnace. The specimens were placed into a retort, and this was carefully closed and sealed. Heat treating was done for 1 hour in a stream of commercial hydrogen. The furnace could be operated at temperatures from 300 to 1000 degrees Cent. (570 to 1830 degrees Fahr.), and the temperature was measured by means of an iron-constantan thermocouple.

The cold rolling experiments were made by means of a small experimental square wire rolling mill. The powder compacts were either molded into suitable bar shapes of $\frac{3}{8}$ inch square cross section, or, as in the case of the hot-compressed specimens, they were cut into small rectangular pieces of about the same cross section. The specimens were continually rolled down and directly reduced in cross sectional area until the first transversal cracks were plainly visible or until an approximate 97 per cent cold rolling reduction was obtained. The reduction in area was then measured by means of a micrometer.

TEST RESULTS

Microstructure—Figs. 2 to 6 give a number of photomicrographs relating to differently processed copper powder compacts. All specimens were molded at a uniform pressure of 50 tons per square inch. Figs. 2 and 3 represent the structures of two specimens, compacted at room temperature and subsequently sintered for 1 hour at different temperatures. The compact of Fig. 2 was sintered at 600 degrees Cent. (1110 degrees Fahr.). Recrystallization of the particles has already taken place, but there are no distinct signs of grain growth. After sintering at 800 degrees Cent. (1470 degrees Fahr.) (Fig. 3) the metallization of the compact has greatly advanced, and grain growth has progressed to a limited extent. The distinct presence of porosity, particularly in Fig. 2, obscures the microstructure of sintered copper, and accounts for the main difference from ordinary copper.

The influence of compacting temperature on the structure of hot-compressed specimens is shown in Figs. 4 and 5. In the specimen compressed at 300 degrees Cent. (570 degrees Fahr.) (Fig. 4) fine pores in the background are numerous, whereas large holes are very rare. Very slight traces of recrystallization in the large particles may be seen in the compact pressed at 300 degrees Cent. (570

degrees Fahr.), but they are much more outstanding after compression at 500 degrees Cent. (930 degrees Fahr.) (Fig. 5). Also at this temperature all cavities have been closed, and the specimen has reached complete density.

The hot-compressed material shows more impurities (especially cuprous oxide) than the hydrogen-sintered compacts.

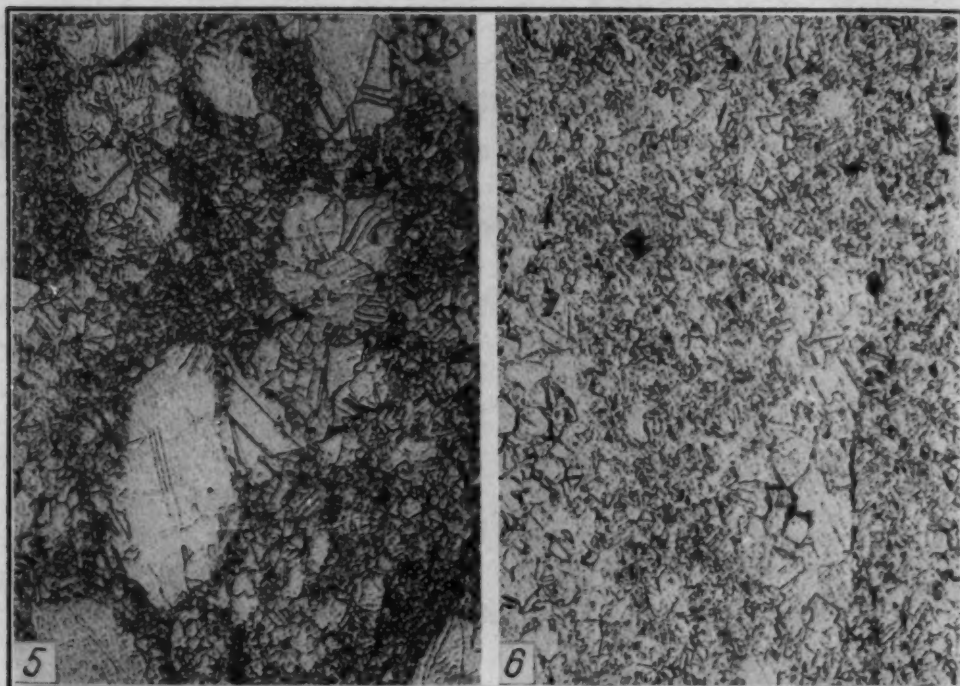


Fig. 5—Same Powder Hot-Compacted at 50 Tons per Square Inch and 500 Degrees Cent. $\times 200$.

Fig. 6—Same Powder Hot-Compacted at 50 Tons per Square Inch and 400 Degrees Cent. Annealed at 800 Degrees Cent. for 1 Hour. $\times 200$.

A specimen of the same powder, hot-compressed at 400 degrees Cent. (750 degrees Fahr.) and 50 tons per square inch was annealed at 800 degrees Cent. (1470 degrees Fahr.) for 1 hour in hydrogen. Its microstructure is shown in Fig. 6.

The difference in the character of microstructures before and after annealing is very marked. A complete crystalline transformation has taken place during the annealing, yielding a fine crystalline, homogeneous structure.

Density—Tests were made to determine the density of hot-compressed and sintered copper powder compacts. Cylindrical slugs, $\frac{5}{8}$ inch in diameter and $\frac{3}{8}$ inch high were used as test pieces. The weight-volume method was used for the tests.

The density of hot-compressed copper powder compacts is plotted against pressure for compacting temperatures of 300 and 500 degrees Cent. (570 and 930 degrees Fahr.) in Fig. 7. Similar graphs are

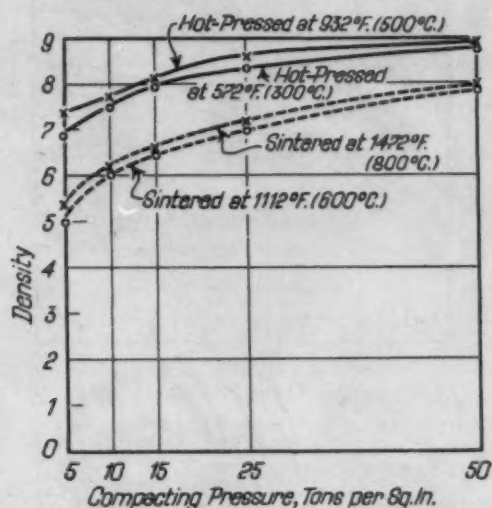


Fig. 7 — Density Versus Compacting Pressure.

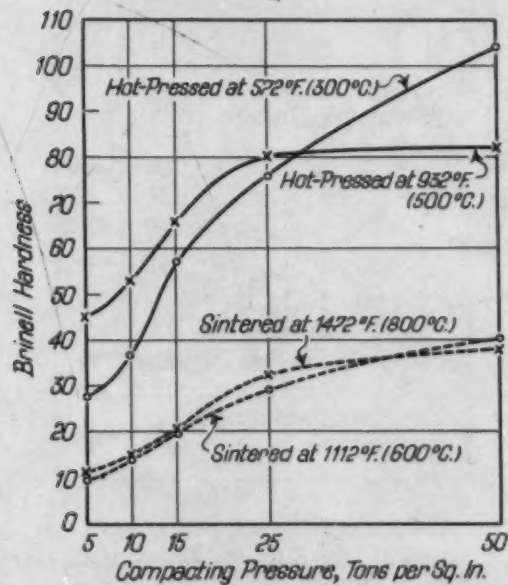


Fig. 8 — Hardness Versus Compacting Pressure.

given in the same diagram for cold-compressed compacts subsequently sintered at 600 and 800 degrees Cent. (1110 and 1470 degrees Fahr.). All curves have a positive slope and are concave upwards. The density value of cast "OFHC" copper (4) is approached by the curve for specimens compacted at 300 degrees Cent. (570 degrees Fahr.)

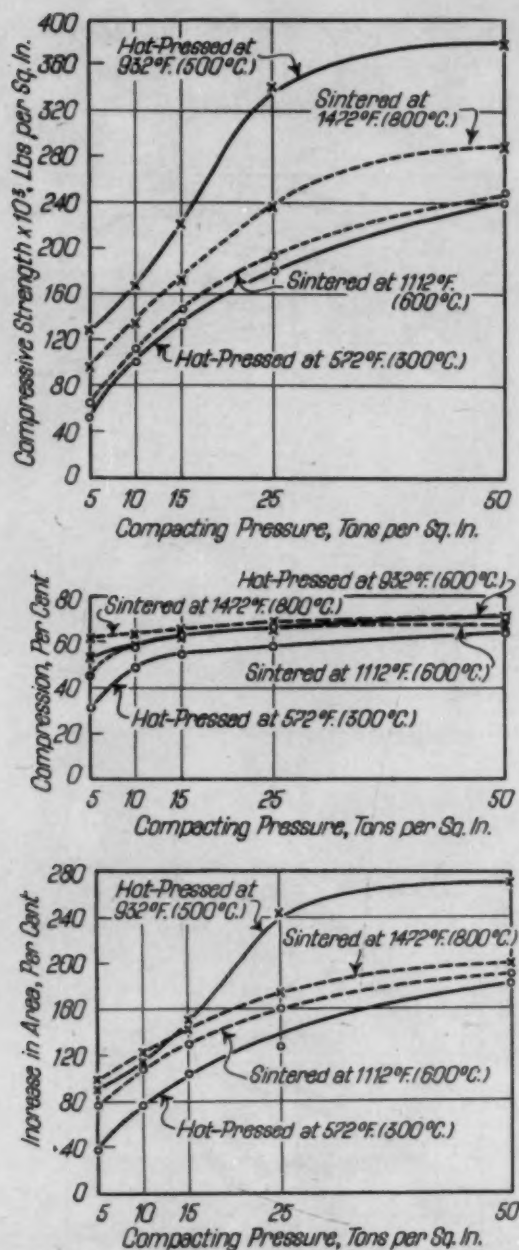


Fig. 9—Compressive Properties Versus Compacting Pressure.

at 50 tons per square inch pressure; it is reached by the curve for specimens hot-pressed at 500 degrees Cent. (930 degrees Fahr.) and 50 tons per square inch pressure. The density curves for the sintered compacts are well below the curves for the hot-pressed material. The increase of sintering temperature from 600 to 800 degrees Cent. (1110 to 1470 degrees Fahr.) raises the points only slightly.

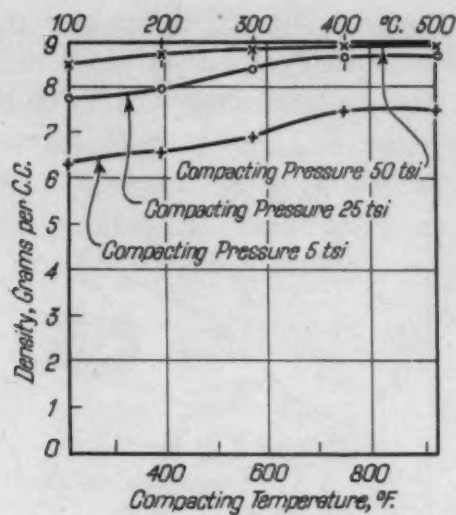


Fig. 10 — Density Versus Compacting Temperature.

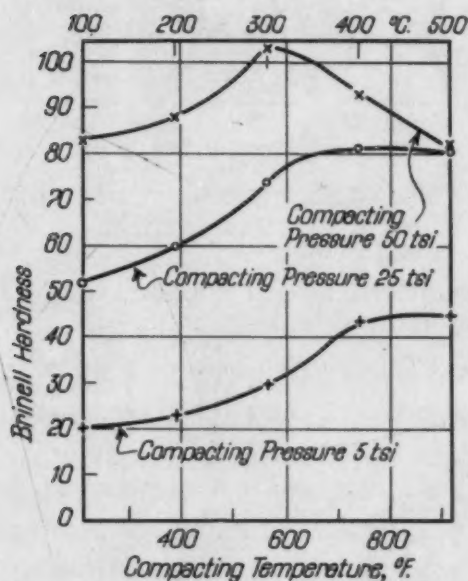


Fig. 11—Hardness Versus Compacting Temperature.

In Fig. 10 the density of hot-compressed compacts is plotted against pressing temperature for various pressures. Here again the curves slope upward, but they are somewhat "s"-shaped. The location of the curves is in accordance with the compacting pressure and for a pressure of 50 tons per square inch the highest curve is obtained. The density value of cast "OFHC" copper is reached at temperatures above 400 degrees Cent. (750 degrees Fahr.) for the 50 tons per square inch curve.

Cold Rolling Property—In Fig. 12 reduction in cross sectional area by cold rolling is plotted as a function of the molding pressure for hot-compressed and sintered compacts. All the curves show a regular increase in reduction of area with increase of compacting pressure. The rolling property of the hot-pressed material is mark-

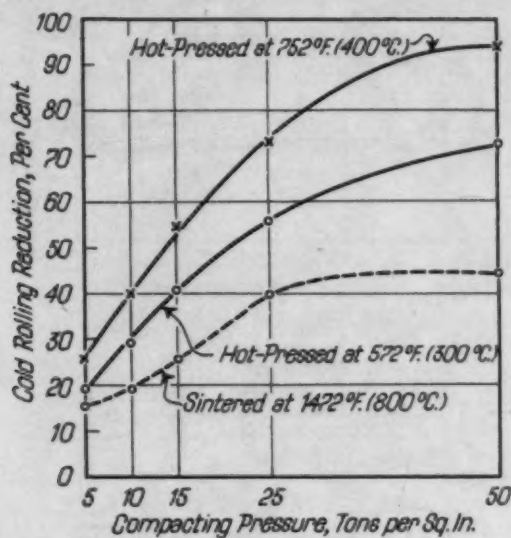


Fig. 12—Rolling Property Versus Compacting Pressure.

edly superior, in comparison with copper, sintered at 800 degrees Cent. (1470 degrees Fahr.), but only the curve for compacts hot-pressed at 400 degrees Cent. (750 degrees Fahr.) approach the value of cast "OFHC" copper, i.e., only these compacts could be directly rolled into wires.

Fig. 13 represents a diagram in which reduction in cross section by cold rolling is plotted against pressing temperature for hot-pressed specimens. All curves rise more or less uniformly with increasing pressing temperatures. The higher the initial pressure the greater is the slope of the curves. The order and regularity of the curves is similar to those of the previous diagram for density (Fig. 10).

Hardness—Hardness tests were made on hot-compressed copper powder compacts. The same cylindrical specimens, previously used for the density tests, were also employed for these hardness tests. The specimens were tested with a Rockwell tester, using $\frac{1}{8}$ -inch ball and 100-kilogram load (Rockwell "E"). This hardness number was then converted into Brinell equivalents by means of an U. S. Bureau of Standard's conversion chart. Direct Brinell measurements

were also made occasionally on some specimens to check the conversion tables.

Hardness values of hot-compressed copper powder compacts were plotted against compacting pressure for temperatures of 300 and 500 degrees Cent. (570 and 930 degrees Fahr.) (Fig. 8). Similar curves are given in the same diagram for compacts sintered at 600 and 800 degrees Cent. (1110 and 1470 degrees Fahr.). In all cases

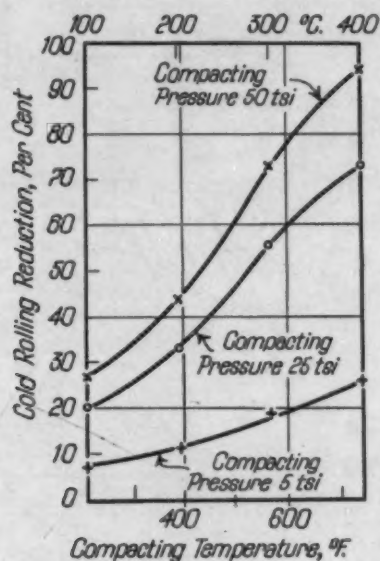


Fig. 13—Rolling Property Versus Compacting Temperature.

the hardness increases with increasing pressures. The maximum hardness values are obtained at 50 tons per square inch for a compression temperature of 300 degrees Cent. (570 degrees Fahr.). However, they do not quite reach the values of Trzebiatowski (3), the maximum is below 110 Brinell. At 500 degrees Cent. (930 degrees Fahr.) the marked bend in the curve indicates the beginning of softening due to recrystallization.

As expected for an annealed material the curves for sintered copper lie far below those for hot-compressed compacts. At 50 tons per square inch they reach the normal hardness of cast copper.

In Fig. 11, the hardness of hot-compressed compacts is plotted as a function of the compression temperature for various pressures. Except for the highest pressure, 50 tons per square inch, the curves again tend to rise continually, although they are slightly "s"-shaped. At 50 tons per square inch compacting pressure, the curves show a

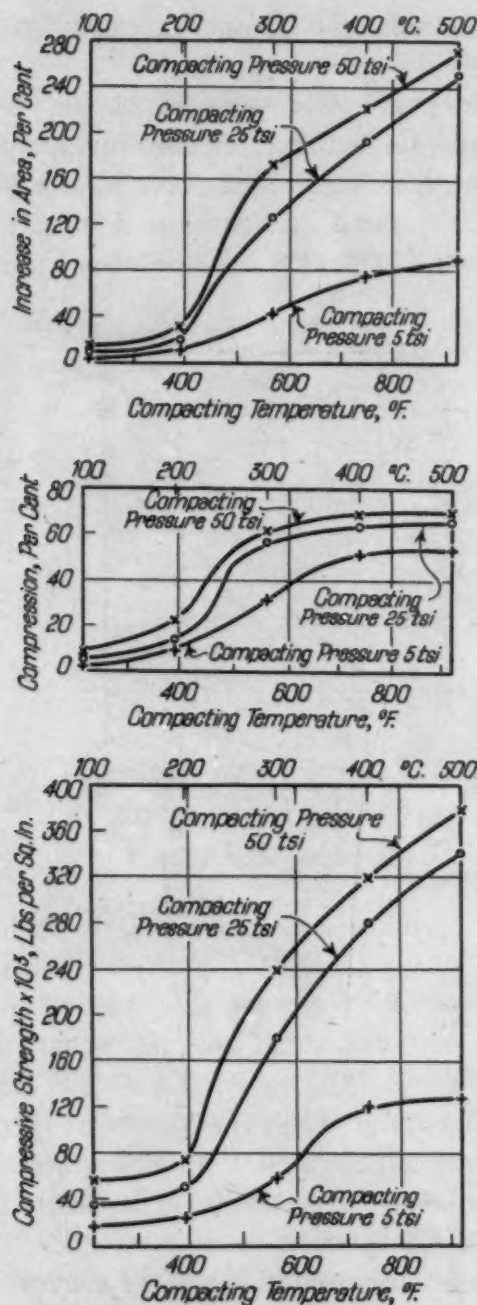


Fig. 14—Compressive Properties Versus Compacting Temperature.

distinct maximum for a molding temperature of 300 degrees Cent. (570 degrees Fahr.). Beyond this temperature hardness is dropping steadily.

Compressive Properties—Compression tests were made on hot-compressed and on sintered compacts. The specimens had an original

diameter of only $\frac{5}{8}$ inch, and did not correspond to the standard specification; but the ratio between length and diameter was about 0.9. The compression tests were made with a 200,000-pound mechanical testing machine. The maximum load which the specimens could take was used as the compressive strength. The change in length and diameter of the samples during testing was determined by micrometer measurements.

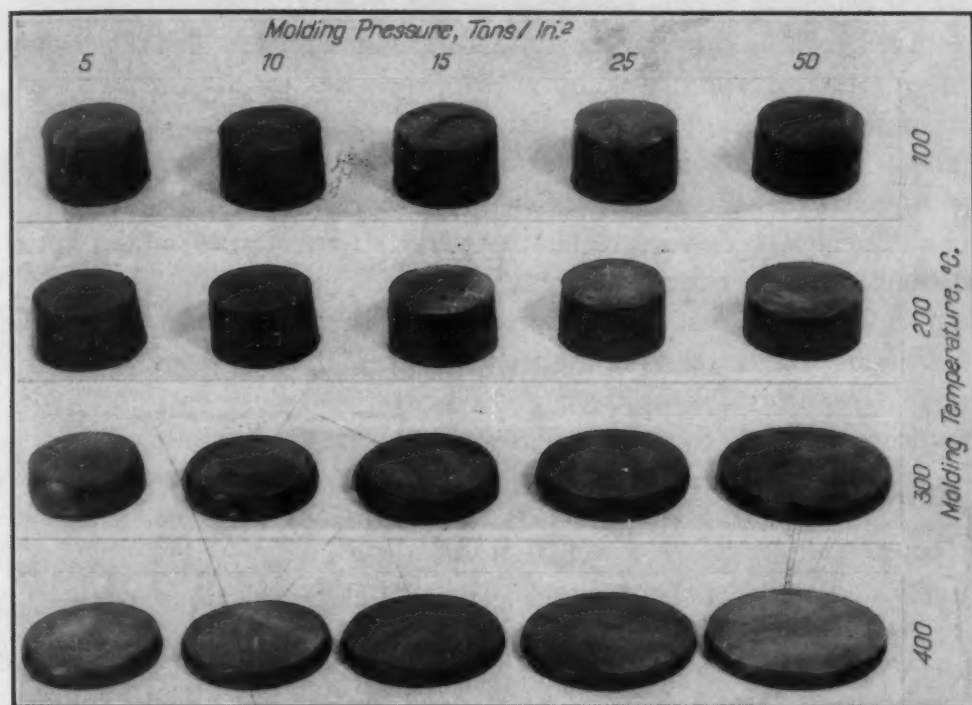


Fig. 15—Tested Compression Specimens from Hot-Compressed Compacts from Electrolytic Copper Powder, Compressed at Different Temperatures and Pressures.

Fig. 15 is a photograph of copper powder compacts prepared by hot compression under different conditions, after compression testing. The influence of pressure or compression temperature is clearly visible. Malleability becomes pronounced only at a pressing temperature of 300 degrees Cent. (570 degrees Fahr.), at which temperature recrystallization of the larger powder particles has begun.

In Fig. 9 plots of the compressive properties of hot-compressed and sintered samples as a function of pressure are shown. Each curve refers to a particular pressing or sintering temperature. All curves have a definitely convex downward shape, and they rise considerably with increasing pressures. An arrangement of the curves as a function of processing method and temperature is recognizable.

The curves for the sintered material lie between those for the compacts hot-compressed at 300 and 500 degrees Cent. (570 and 930 degrees Fahr.). The compressive properties of cast "OFHC" copper of average grain size (4) are equalled or surpassed only by compacts hot-compressed at 500 degrees Cent. (930 degrees Fahr.) and 50 tons per square inch pressure.

In Fig. 14, graphs of the compressive properties of hot-compressed compacts versus pressing temperatures, at constant pressures are given. Each curve refers to a certain pressure. There is generally a tendency for the compressive properties to increase with increasing compression temperatures, up to 500 degrees Cent. (930 degrees Fahr.), and the curves are the higher as the compacting pressure increases. All curves again have an "s"-shaped character, namely a concave downward shape at lower temperatures and a convex shape at higher temperatures.

The data for cast "OFHC" copper are reached or surpassed if the molding was done at the highest pressure, 50 tons per square inch, and temperatures above 400 degrees Cent. (750 degrees Fahr.), or at a pressure of 25 tons per square inch and at the highest temperature, 500 degrees Cent. (930 degrees Fahr.).

SUMMARY

Microscopic tests of hot-compressed compacts indicated that up to a compression temperature of 400 degrees Cent. (750 degrees Fahr.), the structure remains unchanged, except for the beginning of recrystallization in large particles. This confirms Trzebiatowski's statement (3) that the grain growth and marked softening of hot-compressed copper begin at about 500 degrees Cent. (930 degrees Fahr.).

Due to the short time in which hydrogen was allowed to react with the compact in the hot compressing operation, the material generally had more impurities (especially cuprous oxide) than ordinary hydrogen-sintered compacts.

The most distinct difference between the microstructures of hot-pressed and sintered copper was found in the complete absence of cavities in the former material for pressures of 50 tons per square inch and for temperatures above 400 degrees Cent. (750 degrees Fahr.).

Sintered copper resembled more closely the typical polygonal

grain structure of ordinary copper, but disclosed finer grains and cavities in the grain boundaries. At high sintering temperatures, grains showed a tendency to grow and to become more uniform, and cavities diminished and became more evenly distributed.

The ideal value for the density of copper was only reached with hot-compressed compacts, pressed at 50 tons per square inch and above 400 degrees Cent. (750 degrees Fahr.). The density rises with increasing pressure at constant temperatures as well as with increasing temperature at constant pressures. The values for sintered copper varied between 80 and 90 per cent of the corresponding data for hot-pressed compacts.

The rolling reduction curves of the hot-compressed and sintered compacts were very interesting. The values increased with increase of pressure, or pressing temperature. In spite of the fact that the structure of the hot-pressed compacts was still in its initial stage of powder agglomeration without the formation of a uniformly polygonal grain structure, the material offered better rolling properties than the copper sintered as high as 800 degrees Cent. (1470 degrees Fahr.). Only powder compacts hot-pressed at 50 tons per square inch and 400 degrees Cent. (750 degrees Fahr.) could be directly reduced to wire size as in the case of cast copper.

Hot-compressed compacts show a considerable increase in hardness as compared with cold-compressed and sintered materials. Compacts molded at 300 or 400 degrees Cent. (570 or 750 degrees Fahr.) using only 5 tons per square inch pressure have the same hardness as sintered compacts initially compressed to 50 tons per square inch at room temperature, while similar powder compacts compressed at 400 degrees Cent. (750 degrees Fahr.) and 50 tons per square inch have three times the Brinell hardness of cast or sintered copper.

In the case of hot-compressed compacts, hardness rises with increasing pressure as well as with increasing temperature, up to 300 degrees Cent. (570 degrees Fahr.). At higher temperatures, a tendency to soften becomes evident.

In spite of the already mentioned fact that hot-compressed compacts still have the structure of a densely compacted agglomerate of powder particles, the compressive properties of these compacts, if molded at a sufficiently high pressure and temperature, are of considerable significance. They even reach or surpass average values of cast "OFHC" copper. The malleability of the hot-molded compacts, as shown by the compressive strength tests, increases gradually with

increasing temperature, and also, but less markedly, with increasing pressure. Maximum values surpassing those of sintered copper are obtained after recrystallization of the particles has commenced, and after the compacts have become almost perfectly dense.

CONCLUSIONS

In order to obtain maximum values of hardness and strength, the copper powder must be compacted at high pressures. The tendency to entrap air sets a limit to the compacting pressure and porosity due to this can be reduced only if compression of the powders is carried out at elevated temperatures, thus taking advantage of the increase of plasticity of the material. Such temperatures are best selected in the neighborhood of the recrystallization temperature range.

The density of the hot-compressed specimens rises with an increase of pressure or temperature to values far exceeding any data obtained with cold-compacted and subsequently sintered powders.

Hot-compressed compacts also show marked improvement in hardness in comparison with sintered copper. At compression temperatures in the temperature range of recrystallization, e.g., around 300 degrees Cent. (570 degrees Fahr.), the values of hardness obtained for moderate pressures are close to those of cold-compressed and repressed material to which extremely high pressures have been applied. Although in this investigation the high hardness values of Trzebiatowski's research could not quite be reached, probably because he used a higher compacting pressure and a powder much finer than the one used in this work, values of above 100 Brinell as obtained in this investigation, are still of remarkable magnitude for copper.

In spite of the incompletely developed polygonal structure, hot-pressed copper powder compacted at high pressures near the temperature range of recrystallization is malleable enough to be rolled into wire without any intermediate annealing. A certain extent of recrystallization of the larger particles and the lack of porosity may explain such remarkable malleability and ductility of a material, which shows only incipient grain growth. Sintered compacts, on the other hand, cannot be directly reduced to wire size by an ordinary cold compacting and sintering procedure involving short periods of heat treating time, e.g., 1 hour at temperature. It seems likely that

hot-compressed compacts manufactured at even higher temperatures, such as 500 or 600 degrees Cent. (930 or 1110 degrees Fahr.) would yield a material of excellent malleability and ductility, and would compare well to "gross" copper in this respect.

Compressive strength figures of hot-compressed compacts are in definite agreement with observations made during the rolling tests. Evidently high malleability and compressibility are both closely connected with recrystallization and incipient grain growth of the particles, as established in compacts compressed at or slightly above recrystallization temperature. For hot-compressed copper compressive strength values reached figures which can scarcely be met by any cold compacted and sintered material, or even cast material.

The results of this paper on pure copper are of special interest with regard to the powder metallurgy of copper alloys, a field of great commercial importance, where systematic data of the effect of various processing methods are all but scant.

ACKNOWLEDGMENTS

The author wishes to express his appreciation to the American Electro Metal Corporation for the permission to publish the results of this investigation.

Bibliography

1. F. Sauerwald, "Einige neue Versuche zur Herstellung synthetischer Koerper aus Metallpulvern," *Z. Metallkunde*, 1929, p. 22.
2. W. D. Jones, "Principles of Powder Metallurgy," London, Edw. Arnold & Co., 1937, p. 96.
3. W. Trzebiatowski, "Ueber Warmpressversuche an hochdispersen Metallpulvern," *Z. Phys. Chem.*, 1934, A 169, p. 91
4. C. G. Goetzl, "Some Properties of Oxygen-Free High Conductivity Copper," *Metal Industry*, Vol. LIII, 1938, 26, p. 605; *TRANSACTIONS, American Society for Metals*, Vol. 27, 1939, p. 458.

DISCUSSION

Written Discussion: By F. V. Lenel, research metallurgist, Moraine Products Division, General Motors Corp., Dayton, Ohio.

This paper gives us valuable information in a field where hardly any published data exist, the hot pressing of metal powders. Of particular interest are the author's conclusions that hot pressing at temperatures considerably below the usual sintering temperatures of cold-compacted powder gives better physical properties than cold compacting and sintering. It is to be hoped that similar data for iron and iron alloys will also become available as such data

would be of considerably greater practical importance than those on copper powder. Articles made of copper alloys by powder metallurgical methods, such as self-lubricating bearings, or electric motor brushes, are usually used for purposes where high physical properties are of secondary importance compared with other properties. As far as I am aware, articles from pure copper powder have not been used extensively in industry. Such articles seem to be most promising in the electrical industry. It would, therefore, be of interest to know what the electrical conductivity of hot-compressed copper powder compacts is. It is known that the electrical conductivity of cold-compressed and sintered compacts is between 15 and 50 per cent below the conductivity of wrought copper which is not surprising in view of the low density of the compacts. The author shows that with the correct combination of pressure and temperature he can get densities similar to wrought material for his hot-pressed compacts. One would, therefore, expect that the electrical conductivity would also approach that of wrought material unless the oxide inclusions which the author finds in hot-compressed materials interfere with the conductivity.

I should like to add a few more general remarks which are concerned not only with the paper under discussion but with the general problem of determining the properties of metal powder compacts. This is the problem which is peculiar to powder metallurgy and distinguishes its products from cast or wrought metal products. The properties of powder compacts obviously depend on the compacting pressure and the conditions of heat treatment. They also depend on the properties of the powder used. The author in his paper gives us the chemical and the screen size analysis of the powder he uses. Somebody not well acquainted with the intricacies of powder metallurgy may well assume that the powder is hereby well identified and that any powder with the same chemical and screen size analysis will give the same physical properties when compacted under a given test condition. But this is not so. Even powders which are made by the same general methods, let us say, all electrolytic copper powder or all reduced copper powder of a given chemical and screen analysis, will not necessarily give the same properties when compacted and sintered.

Unfortunately, we do not know what all the factors are which determine the properties of powders. We know in a general way that the shape of powder particles, the degree of their porosity, the particle size distribution in sizes smaller than 325 mesh, influence these properties of powder compacts. We also know that changes in the manufacturing technique of the powders, for instance the temperature of reduction for reduced powders, the current density, or concentration of the electrolyte for electrolytically produced powder, the amount of cold work applied in disintegrating the powder cake, etc., may greatly influence those properties.

But we really ought to know exactly what properties of a given powder influence the properties of a compact made from it and to what degree changes in powder properties are reflected in changes in properties of the compacts. Until these problems are investigated, numerical values for the properties of compacts will be of value only inasmuch as they show trends and general relationships, not as absolute values. I believe that such fundamental investigations upon the influence of powder properties or properties of compacts are

not only of great scientific interest but also of vital practical importance to both powder manufacturers and users.

Written Discussion: By W. H. Swanger, metallurgist, National Bureau of Standards, Washington, D. C.

This paper presents much useful information on certain physical properties that can be obtained in compacted copper powder, but unfortunately gives no original information on the strength properties that might be of most interest to users, namely results of tensile, impact or bending tests. The author's compression tests indicate that the compacts can be made with excellent malleability, but an entirely erroneous conception of strength is implied by the values given in Figs. 9 and 14 for compressive strength.

The author does not state the criterion he used in his compression tests to determine "the maximum load which the specimens could take —". The photographs of the tested compression specimens (Fig. 15) show no indication of the shattering fractures obtained in brittle materials such as cast iron, concrete, etc. It is well known that very malleable metals can be squeezed between the platens of a testing machine until they carry the highest load that the machine can apply. For such materials "compressive strength" has no meaning unless accompanied by a statement of the permissible amount of distortion, which may be chosen arbitrarily, but to be useful, must be in accordance with engineering requirements.

The compressive properties of compacted and sintered copper powder, as well as those of other compacted powders, are of interest, but the "compressive strength" should be determined as the stress required to produce a permanent set (deformation limit) in compression, of a magnitude in keeping with engineering requirements. See American Society for Testing Materials Specifications E9-33T, p. 1237, and B22-38T, p. 1072, American Society for Testing Materials Standards, Part I, Metals, 1939.

It is to be hoped that the author will continue his investigations of compacted metal powders to furnish the much needed information on their strength properties.

Written Discussion: By John Wulff, associate professor, Massachusetts Institute of Technology, Cambridge, Mass.

Dr. Goetzel's data on hot-pressed compacts supply a welcome addition to the literature on powder metallurgy. Of practical import are the observations that hot-pressed compacts have higher density and hardness and permit greater degrees of cold deformation than cold-pressed and sintered compacts. Whether the malleability of these compacts may be attributed to increased density or to other factors is of interest to all who work with metal powders. In this respect Kelley, in a paper at the last meeting of the American Institute of Mining and Metallurgical Engineers, showed that by sintering cold-pressed compacts of iron and other metals at sufficiently high temperatures bulk densities could be achieved.

Has Dr. Goetzel tested the malleability of copper or other metals sintered at higher temperatures than reported in this paper?

We hope that Dr. Goetzel will continue his fact finding and publication. Information on the properties of hot-pressed iron and steel compacts would

be of great interest. Preliminary unpublished work of various workers in the field suggests limited applicability due primarily to low production rates and high die costs.

Oral Discussion

D. C. JILLSON:¹ My question is in line with that of Professor Wulff. The author has compared the results of physical tests on specimens cold-pressed and sintered at temperatures of 600 and 800 degrees Cent. with those of specimens hot-pressed. The results, he said, indicated certain advantages for the hot pressing method. However, a temperature of 800 degrees Cent. is considerably below the melting point of copper. The author did not indicate that he had determined the optimum sintering temperature for cold-pressed copper specimens.

I have in mind some work that was done on cold-pressed iron powders,² where a maximum in certain physical properties occurred on sintering in the neighborhood of 1000 degrees Cent. Going to slightly higher temperatures, these properties decreased, but further work showed that on continuing to still higher temperatures, almost up to the melting point of iron, there was a very sharp change in the properties of the resulting compacts. There was a very steep increase in certain properties, and another maximum, much higher than the first, was obtained.

I think that we should be careful, in comparing results of different methods, to determine truly optimum conditions for each, and remember that there may be more than one maximum.

Author's Reply

Dr. Lenel has brought up a most interesting point in stressing the many general factors which influence the properties of metal powder compacts. In order to investigate all the factors as mentioned by Dr. Lenel, a thorough and extensive research program would have to be carried out. For this reason this paper is only thought of as a beginning in the direction of a complete investigation of all the important factors influencing the material. Additional research is planned. On the other hand, it may be said that some of the factors mentioned have, evidently, surprising as it is, a less distinct bearing, if any, on hot-pressed material than they have on conventional cold-pressed and sintered compacts. Professor W. D. Jones in a recent series of publications has shown that the hot pressing methods produce copper alloy compacts whose properties are rather independent of certain factors, such as particle size distribution, oxide films around the particles, type of atmosphere used during hot pressing and others. This is remarkable if one considers that all these factors influence sintered metals to a great extent. Dr. Lenel's statement that the main interest of powder metallurgists today is focused on the behavior of sintered and hot-pressed ferrous materials rather than on nonferrous metals can also be supported wholeheartedly. Research on these lines has already

¹Investigator, research division, New Jersey Zinc Co., Palmerton, Pa.

²Private communication from Professor Claire Balke, Stevens Institute of Technology, Hoboken, N. J.

been started, but it is, as yet, too early to make definite statements about the trends and specific outcome of the tests.

Mr. Swanger also mentioned an interesting point by discussing the testing method that the compacts of the present paper were subjected to. Tensile, impact conductivity and bending tests require long and rectangular specimens, which at the present are still very difficult to be molded at elevated temperatures. The entire problem of die design for hot pressing is very involved and only in its early stages. On the other hand, for cylindrical specimens a die is not quite as difficult to construct and to bring into a working condition. Since compressive strength tests are made with cylindrical specimens, this type of test was thought of as some kind of indication of the specimens' malleability rather than strength since malleability is probably more significant for powder compacts than strength. Mr. Swanger's point that the "maximum load" figure does not mean much in conventional compressive strength testing is correct. However, it must be borne in mind that in the case of powder compacts we often deal with very brittle materials where a definite set of a deformation limit may not be reached at all with some of the specimens of a test series. For this reason the "maximum load" figure appeared to be the best solution under the given circumstances. In the tests of this investigation as "maximum load" the point was taken at which the beam of the tension machine dropped for the first time at a constant load. Thereafter the load was released quickly so that a crushing of the brittle specimens was prevented in most cases. However, all specimens after testing showed the well-known conical shattering fractures, although, evidently, the picture of Fig. 15 does not show such features clearly.

May I thank Professor Wulff for his encouraging criticism and may I state that I fully agree that the densities of copper sintered at a higher temperature are higher, and, accordingly, properties might go up. The maximum temperature which I used in my tests was 900 degrees Cent., 100 degrees higher than shown in these tests. I found that at this temperature a severe shrinkage and distortion is taking place. Besides I purposely selected the temperature of 800 degrees Cent. because the commercial use of that temperature was well in mind. Porous bronze bearings are generally sintered in the neighborhood of 800 to 825 degrees Cent. A much higher temperature will result in melting these materials, and, since this paper was considered as a more basic paper, it was decided to stay in the same temperature range as the copper component of the bearings.

Besides, it should be considered that even if one were to deal with straight copper, the commercial point is also important, and heating a furnace permanently at 1000 degrees Cent. is more expensive than heating it at 800 degrees Cent. As far as known there are several products made from straight copper powders for electrical purposes, and all these are sintered commercially at a temperature in the neighborhood of 800 degrees Cent.

The other point brought up by Professor Wulff is that experimenting with iron would be more desirable than with copper. As already stated above, I am fully in agreement with such suggestion, but there are certain difficulties yet to be overcome. It is quite difficult to hot press from a certain temperature

limit at high pressures. Whereas a temperature of 400 or 500 degrees Cent. is sufficient to make copper completely dense, this is not enough to make iron dense and strong. Right now I am not prepared to make any further statements concerning this matter.

With regard to the much appreciated discussion of Mr. Jillson, I can only repeat that higher temperatures would result in severe dimensional changes of the specimens during sintering, and for this reason high temperature sintering tests were left out in this particular research.

I agree with Mr. Jillson that one has to be careful in making comparisons. This, however, also is true for comparing copper with iron. In iron we have an entirely different material, with a higher melting point, with more rigid powder particles and with transformation points at elevated temperatures. I am not familiar with tests made on iron where two property maxima were found at different temperatures. The type of the initial powder has a very important bearing on the final physical properties, as was just emphasized by Dr. Lenel, and, in addition, it might have been possible that the transition of the material through the transformation temperature had some effect on these test results.

In conclusion it may be said that it is planned to establish tensile strength, bending and impact data on the same material treated in this investigation as soon as proper molding facilities can be made available.

THE PRECIPITATION REACTION IN COLD-ROLLED PHOSPHOR BRONZE: ITS EFFECTS ON HARDNESS, CONDUCTIVITY, AND TENSILE PROPERTIES

BY R. H. HARRINGTON AND R. G. THOMPSON

Abstract

Since the discovery that long-time heating of cold-rolled phosphorus-bronze results in an increase in electrical conductivity, etc., there has been well founded suggestion that the copper-tin equilibrium diagram should be altered to account for precipitation. The present paper proposes: (1) that the conventional copper-tin diagram should show room temperature solid solubility at about 16 per cent tin; (2) that the precipitation effect, found upon heating strained material, should be indicated along a third axis of strain (or "pressure"), perpendicular to the temperature and composition axes at their intersection. The principles governing the temperature-hardness curve of recrystallization of these cold-worked solid solutions are defined, and it is suggested that these principles apply in degree to all cold-worked solid solutions. Data are given to show that the elastic limit, proportional limit, elongation, conductivity, and fatigue properties are advantageously affected by precipitation effects in localized "strain domains."

INTRODUCTION

OF several exemplary references (1), (2), (3), (4),¹ Dahl and Haase (1) give a concise description of the effects to be discussed and amplified. The facts are briefly as follows:

1. The electrical conductivity of cold-worked bronzes is increased by heating for long periods of time below the recrystallization temperature.

2. The effect is measurable in binary bronzes with as little as 3 per cent tin content. On introducing additional elements, this lower limit is decreased to about 1 per cent tin.

3. The effect is not discoverable in thoroughly annealed bronzes when reheated for even extreme periods of time at the lower "tempering" temperatures.

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented before the Twenty-second Annual Convention of the Society held in Cleveland, October 21 to 25, 1940. Of the authors, R. H. Harrington is research metallurgist, and R. G. Thompson is metallurgist, Works Laboratory, General Electric Company, Schenectady, N. Y. Manuscript received July 5, 1940.

4. The effect increases in degree with increase in tin content and with increase in the amount of plastic deformation. Thus an increase in electrical conductivity of about 40 per cent may be induced by heating 90 per cent cold-rolled 13.5 tin bronze for 1000 hours at 250 degrees Cent. (480 degrees Fahr.).

It therefore becomes of interest to evaluate this apparent precipitation effect in terms of the practical application of commercial material as well as to add to the available research data.

MATERIAL

A commercial bronze of the 8 per cent tin—0.1 per cent phosphorus type was the basis for this study. The material, in the form of 20 mil strip, 1 inch wide, was tested in three degrees of cold work: "6 numbers hard" (50 per cent cold reduction), "10 numbers hard" (68 per cent cold reduction), and "20 numbers hard" (90 per cent cold reduction). The effect of grain size was determined for the 6 and 10 Numbers Hard grades.

TESTS

The tests resolved themselves into the following successive steps:

1. Determination of the temperature-hardness recrystallization curves. Separate samples of each grade of material were heated for 4 hours at each selected temperature. Hardness measurements were made with the Rockwell superficial tester, using the 30-T Scale. Ten or more readings were made, and averages were taken.

2. Determination of the maximum "tempering temperature"² without undue loss of hardness in the cold-rolled materials for extended periods of time.

3. Effect of strain and "tempering temperature" on the electrical conductivity as compared to conductivity of material as received in the cold-rolled state and after complete recrystallization.

4. Effect of the induced precipitation reaction on the tensile properties. The proportional limits were measured by means of a Sayre extensometer with an accuracy of 1 part in 100,000. The elastic limits were determined with the Sayre extensometer as the stress required to produce a permanent elongation of 1 to 2 parts in 100,000.

²These "tempering temperatures" prove to be aging temperatures

5. Effect of the induced precipitation reaction on the endurance properties. The S-N curves were determined for material cold-rolled 8 numbers hard (60 per cent cold reduction), using the well known strip-fatigue testing machine developed by the Bell Telephone Laboratories.

6. Photomicrographs of the material in the various conditions.

TEMPERATURE-HARDNESS RECRYSTALLIZATION CURVES

Some new principles affecting recrystallization have been suggested by one of the present authors in a previous report (4).

There are two types of strain to be considered in a cold-worked solid solution alloy: (a) Elastic strain and (b) Plastic strain. By

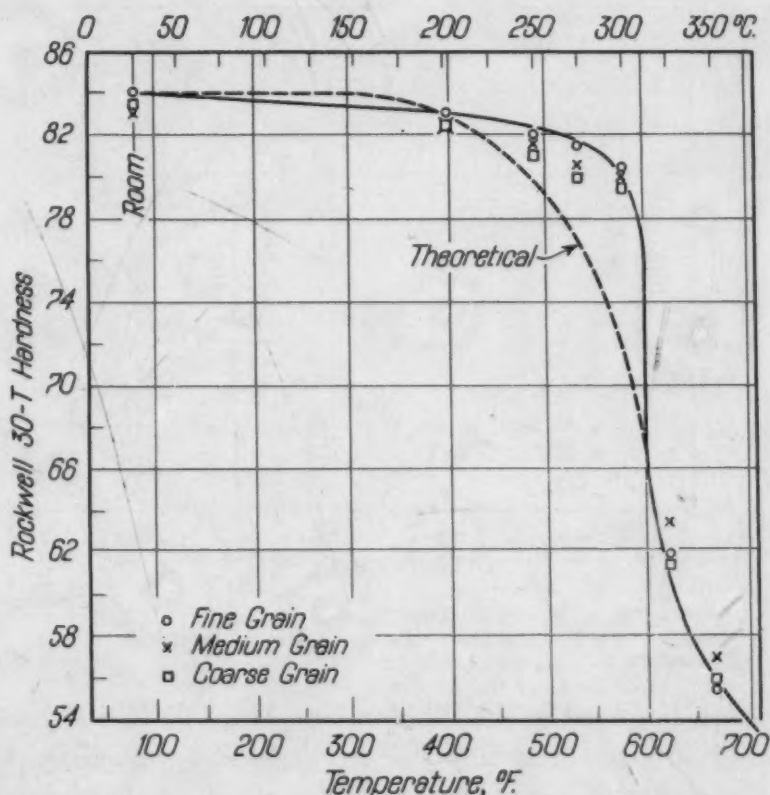


Fig. 1—8 Per Cent Tin Phosphor Bronze. Cold-Rolled 6 Numbers Hard. Hardness Values After 4 Hours at Stated Temperatures.

“elastic strain” is meant the lattice distortion which is relieved by restoration diffusion. By “plastic strain” is meant the lattice distortion which is relieved only by recrystallization. Plastic deformation results in the occurrence of both types of strain. Thus in cold-worked

metals or solid solution alloys, heating for a relatively short time at temperatures below the "recrystallization temperature" will effect relief from elastic strain with little or no effect on the plastic strain.

The temperature-hardness recrystallization data are plotted for the 6-, 10-, and 20 Numbers Hard grades in Figs. 1, 2, and 3 respectively. No practical differences can be discerned for the different grain sizes for either grade of the 6 and 10 Numbers Hard. Since

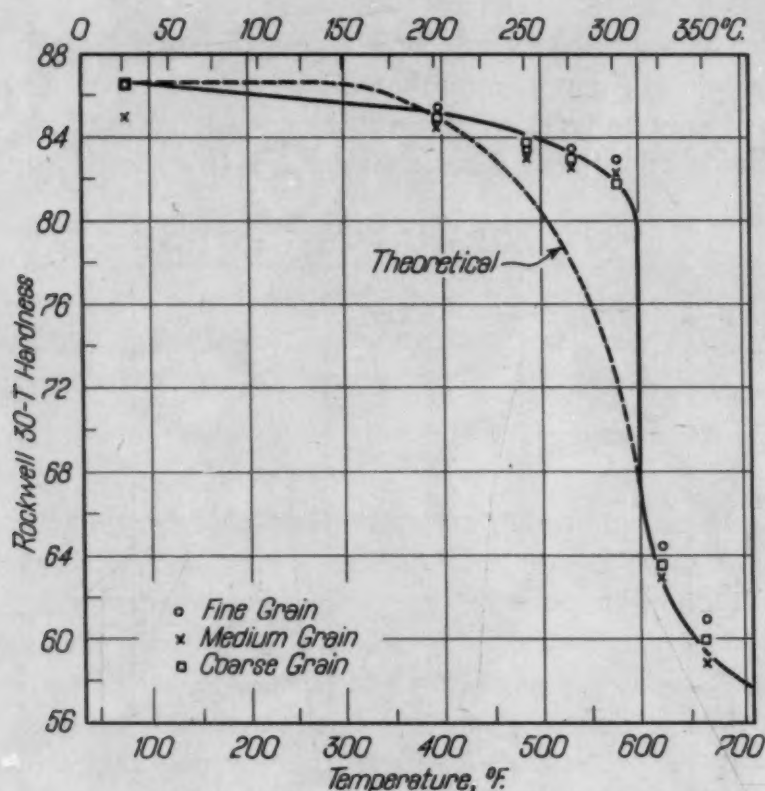


Fig. 2—8 Per Cent Tin Phosphor Bronze. Cold-Rolled 10 Numbers Hard. Hardness Values After 4 Hours at Stated Temperatures.

the reduction involved in the 20 Numbers Hard condition is so great, only one grain size, Medium, was studied. As previously reported (4), if the recrystallization is simply the relief of the plastic strain, the temperature-hardness recrystallization curve should be of the probability type with normal distribution of local areas according to the degree of strain. Thus those few areas of the highest plastic strain will recrystallize at lowest temperature, etc. The resulting temperature-hardness relationship is indicated by the "Theoretical" curves (broken line) drawn in Figs. 1, 2, and 3. The relation of

these theoretical curves to the actual data curves will be discussed in detail for the 20 Numbers Hard grade (Fig. 3) which gives the maximum effect of the induced precipitation reaction.

As an introduction to the discussion of Fig. 3, it will be recalled that a previous report (2) shows for these cold-rolled bronzes that (a), with short times at temperatures below the recrystallization temperature, the electrical conductivity is measurably decreased, indica-

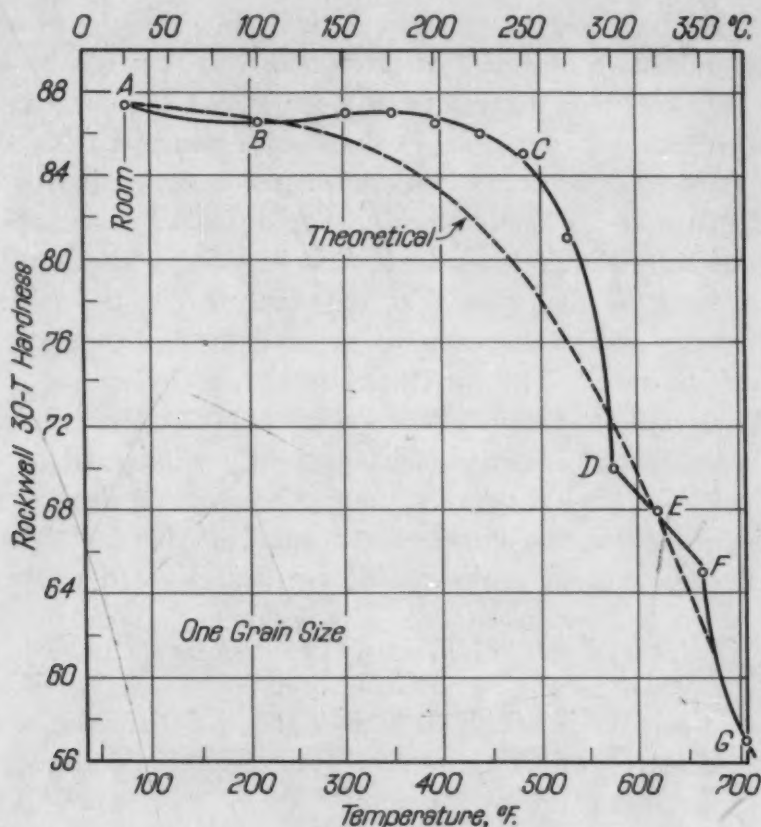


Fig. 3—8 Per Cent Tin Phosphor Bronze. Cold-Rolled 20 Numbers Hard. Hardness Values After 4 Hours at Stated Temperatures.

tive of additional internal strain when the contrary should be expected with normal relief of the strain induced by the cold work; that, (b), with extended periods of time at the higher temperatures, but still below the recrystallization temperature, the electrical conductivity is markedly increased to values above the electrical conductivity for the annealed material. This is one of the chief criteria for the corresponding stages of precipitation: (a) aggregation of solute atoms within the solvent lattice and (b) true precipitation.

A. Theoretical Recrystallization Curve—The broken line curve represents the recrystallization of cold-worked solid solution bronze were it not susceptible to precipitation when reheated under conditions of internal plastic strain, and neglecting the early relief from elastic strain. Thus, from A to G, recrystallization occurs from plastic strain with no precipitation effects (Fig. 3).

B. Actual Recrystallization Curve—The solid line curve represents the actual data for this specific temperature-hardness recrystallization reaction. The range from A to B shows a slight decrease in hardness, indicative of relief of strain. From B to C the hardness actually increases to a maximum between B and C indicative of precipitation effects. From C to D the hardness drops with abnormal rapidity as re-solution occurs with consequent elimination of the precipitated phase as an agent blocking the diffusion necessary for recrystallization. With release from this blocking effect, the internal energy content of the system at temperatures in the range from C to D causes recrystallization to proceed in this range at an abnormally high rate. This, in turn, results in hysteresis effects at D and F, deviations from the theoretical curve. The effects of the induced precipitation reaction are apparently eliminated at G, and, above about 375 or 400 degrees Cent. (705 or 750 degrees Fahr.), the hardness-temperature curves are normal, the hardness decreasing with the coarsening of grain size at further elevated temperatures.

TEMPERATURE, TIME, HARDNESS EFFECTS

From the data in Table I, it is concluded that the maximum aging temperatures, effecting no recrystallization and relatively small decrease in hardness, are as follows:

For 6 Numbers Hard cold reduction—250 degrees Cent. (480 degrees Fahr.).

For 10 Numbers Hard cold reduction—250 degrees Cent. (480 degrees Fahr.).

For 20 Numbers Hard cold reduction—225 degrees Cent. (435 degrees Fahr.).

The detailed data for the 20 Numbers Hard material, aged at 225 degrees Cent. (435 degrees Fahr.), show an age hardening maximum for 50 to 80 hours at temperature. Increasing the aging temperature to 250 degrees Cent. (480 degrees Fahr.) eliminates the age hardening maximum although the blocking effect of the induced

Table I
Temperature, Time, Hardness Data

6 Numbers Hard			
As-received			83.5 (30-T)
75 hours at 300 degrees Cent. (570 degrees Fahr.)			56.0
75 hours at 275 degrees Cent. (525 degrees Fahr.)			71.0
100 hours at 250 degrees Cent. (480 degrees Fahr.)			78.0
10 Numbers Hard			
As-received			86.0 (30-T)
75 hours at 300 degrees Cent. (570 degrees Fahr.)			60.0
75 hours at 275 degrees Cent. (525 degrees Fahr.)			69.0
100 hours at 250 degrees Cent. (480 degrees Fahr.)			80.0
20 Numbers Hard			
As-received			87.5 (30-T)
4 hours at 250 degrees Cent. (480 degrees Fahr.)			85.0
10 hours at 250 degrees Cent. (480 degrees Fahr.)			84.0
20 hours at 250 degrees Cent. (480 degrees Fahr.)			83.0
50 hours at 250 degrees Cent. (480 degrees Fahr.)			82.5
80 hours at 250 degrees Cent. (480 degrees Fahr.)			82.0
100 hours at 250 degrees Cent. (480 degrees Fahr.)			81.0
4 hours at 225 degrees Cent. (435 degrees Fahr.)			85.0
10 hours at 225 degrees Cent. (435 degrees Fahr.)			84.0
20 hours at 225 degrees Cent. (435 degrees Fahr.)			85.0
50 hours at 225 degrees Cent. (435 degrees Fahr.)			85.5
80 hours at 225 degrees Cent. (435 degrees Fahr.)			85.5
100 hours at 225 degrees Cent. (435 degrees Fahr.)			84.5

Table II
Effect of Strain and Temperature on Electrical Conductivity

Cold-Rolled	Grain Size	Treatment	Electrical Conductivity*
6 Nos.	Medium	As-received	12.2 per cent
		Annealed 15 min. 600 degrees Cent. (1110 degrees Fahr.)	12.6
		4 hours at 250 degrees Cent. (480 degrees Fahr.)	12.7
		10 hours at 250 degrees Cent. (480 degrees Fahr.)	12.7
		100 hours at 250 degrees Cent. (480 degrees Fahr.)	12.9
	Coarse	As-received	12.1 per cent
		Annealed 15 min. 600 degrees Cent. (1110 degrees Fahr.)	12.3
		4 hours at 250 degrees Cent. (480 degrees Fahr.)	12.4
		10 hours at 250 degrees Cent. (480 degrees Fahr.)	12.4
		100 hours at 250 degrees Cent. (480 degrees Fahr.)	12.5
	10 Nos. Fine	As-received	12.6 per cent
		Annealed 15 min. 600 degrees Cent. (1110 degrees Fahr.)	12.9
		4 hours at 250 degrees Cent. (480 degrees Fahr.)	12.9
		10 hours at 250 degrees Cent. (480 degrees Fahr.)	13.1
		100 hours at 250 degrees Cent. (480 degrees Fahr.)	13.3
	Medium	As-received	12.3 per cent
		Annealed 15 min. 600 degrees Cent. (1110 degrees Fahr.)	12.6
		4 hours at 250 degrees Cent. (480 degrees Fahr.)	12.7
		10 hours at 250 degrees Cent. (480 degrees Fahr.)	12.8
		100 hours at 250 degrees Cent. (480 degrees Fahr.)	13.1
	Coarse	As-received	12.1 per cent
		Annealed 15 min. 600 degrees Cent. (1110 degrees Fahr.)	12.4
		4 hours at 250 degrees Cent. (480 degrees Fahr.)	12.5
		10 hours at 250 degrees Cent. (480 degrees Fahr.)	12.7
		100 hours at 250 degrees Cent. (480 degrees Fahr.)	12.8
20 Nos.	Medium	As-received	12.04 per cent
		Annealed 15 min. 600 degrees Cent. (1110 degrees Fahr.)	12.41
		4 hours at 225 degrees Cent. (435 degrees Fahr.)	12.50
		10 hours at 225 degrees Cent. (435 degrees Fahr.)	12.53
		100 hours at 225 degrees Cent. (435 degrees Fahr.)	14.53

*Electrical Conductivity in terms of Cu = 100 per cent. Measured at 25 degrees Cent.

precipitation reaction is almost as effective as with aging at 225 degrees Cent. (435 degrees Fahr.). Similarly, in the cases of lower internal strain in the 6 and 10 Numbers Hard material, aged at 250 degrees Cent. (480 degrees Fahr.), the plastic strain, while it results in the same induced precipitation reaction with consequent blocking of recrystallization, is not sufficient in degree to cause a measurable maximum in age hardness. These observations are also in agreement with the temperature-hardness recrystallization curves in Figs. 1, 2, and 3, for comparison of the three degrees of strain.

EFFECT OF STRAIN AND SUBSEQUENT AGING ON THE ELECTRICAL CONDUCTIVITY

Table II gives the data showing the comparison of the electrical conductivities for the specified conditions: as cold-rolled, annealed at 600 degrees Cent. (1110 degrees Fahr.), and 4, 10, and 100 hours at the specified aging temperatures. With the exception of the 20 Numbers Hard material, aged for 100 hours at 225 degrees Cent. (435 degrees Fahr.), the increases in electrical conductivity were very small although within the range of measuring accuracy. The results are very consistent as referred to: (a) variation in amount of cold work (plastic strain), (b) grain size of the cold-worked material, (c) extension of time of aging at constant temperature. (The appended photomicrographs show complete structural recrystallization for the commercial anneals previous to final cold rolling, and no visible change in structures of the samples given aging treatments.) The sample of the 20 Numbers Hard material, aged 100 hours at 225 degrees Cent. (435 degrees Fahr.), gave a 17 per cent increase in the electrical conductivity over the value for the same stock after annealing.

Comparing the electrical conductivities after the 100-hour aging treatments with the values for the annealed material of each lot, the percentage increases in electrical conductivities are as follows:

- 6 Numbers Hard, Medium Grain—2.4 per cent
- 6 Numbers Hard, Coarse Grain—1.7 per cent
- 10 Numbers Hard, Fine Grain—3.1 per cent
- 10 Numbers Hard, Medium Grain—4.0 per cent
- 10 Numbers Hard, Coarse Grain—3.2 per cent
- 20 Numbers Hard, Medium Grain—17.0 per cent

These increases in electrical conductivities, together with the hardness variations cited above and the X-ray investigations of

previous reports (2), appear to describe definitely an induced precipitation reaction.

EFFECT OF STRAIN AND SUBSEQUENT AGING ON THE TENSILE PROPERTIES

Tables III, IV, and V give the data respectively for the 6, 10, and 20 Numbers Hard materials as cold-rolled, aged for extending

Table III
Effect of Strain and Temperature on Tensile Properties, Material Rolled 6 Nos. Hard

Time at 250 Degrees Cent. (480 Degrees Fahr.)	Grain Size	Prop. Limit	Elastic Limit	Tensile Strength	Per Cent Elongation in 2 Inches	Ex 10 ⁻⁶
0 hours	Medium	52,000	80,500	110,500	5	16.0
4 hours	Medium	52,000	44,000	104,100	22	16.6
10 hours	Medium	51,000	50,400	97,650	22	16.6
100 hours	Medium	50,000	86,600	94,000	26	16.6
0 hours	Coarse	47,250	51,600	108,000	8	15.9
4 hours	Coarse	44,500	48,300	98,000	20	15.9
10 hours	Coarse	45,750	57,000	95,000	21	16.3
100 hours	Coarse	50,000	51,300	92,500	22	16.6

Table IV
Effect of Strain and Temperature on Tensile Properties, Material Rolled 10 Nos. Hard

Time at 250 Degrees Cent. (480 Degrees Fahr.)	Grain Size	Prop. Limit	Elastic Limit	Tensile Strength	Per Cent Elongation in 2 Inches	Ex 10 ⁻⁶
0 hours	Fine	57,000	61,100	124,750	2	16.1
4 hours	Fine	58,000	71,400	108,500	16	16.2
10 hours	Fine	58,000	66,000	107,000	17	16.4
100 hours	Fine	57,000	66,500	98,100	21	16.8
0 hours	Medium	58,250	negative drift	118,200	3	16.2
4 hours	Medium	58,500	66,000	106,200	14	16.3
10 hours	Medium	51,700	64,200	104,500	14.5	16.5
100 hours	Medium	48,000	58,600	98,400	17.5	16.6
0 hours	Coarse	54,750	82,800	126,500	2.5	16.0
4 hours	Coarse	56,750	67,900	107,400	12.5	16.2
10 hours	Coarse	59,250	71,500	104,200	12.5	16.5
100 hours	Coarse	55,000	60,600	101,200	16.0	16.7

Table V
Effect of Strain and Temperature on Tensile Properties, Material Rolled 20 Nos. Hard

Time at 225 Degrees Cent. (435 Degrees Fahr.)	Prop. Limit	Elastic Limit	Tensile Strength	Per Cent Elongation in 2 Inches	Per Cent Electrical Conductivity
0 hours	55,000	? *	136,500	1	12.04
4 hours	61,700	66,600?	124,200	6	12.50
10 hours	58,350	?	121,000	7	12.53
100 hours	63,350	73,300?	117,100	11	14.53
Annealed	16,700		56,700	62.5	12.41

* Too much "negative drift" for certainty.

periods of time at specific temperatures, and, for the 20 Numbers Hard material, for the annealed stock.

General Effects—1. Proportional limits show relatively small anomalies, generally an increase with specific aging treatments.

2. "Elastic limits", really yield strengths at a permanent deformation of 1 part in 100,000, show rather wide anomalies with specific heat treatments resulting usually in no loss in value, and, in some cases, real increases. A case for further interesting research is that of the 20 Numbers Hard material, in which the "negative drift," upon release of stress, was so great as to make it impossible to measure the elastic limit with any real meaning.

3. Tensile strengths decreased at a decreasing rate with extension of time of aging.

4. Elongations increased markedly with extension of time of aging. This is most surprising in view of the degree and direction of change noted in the values for the other tensile properties, especially when the reaction involved definitely appears to be a precipitation.

A reasonable explanation for this increase in elongation, together with the observed (and usually advantageous) anomalies in proportional and elastic limits, seems to be that of strain-domain-precipitation within the individual grains. The domains of strain-induced precipitation would act to maintain or increase the proportional limits, elastic limits, and age hardness while the trans-grain paths of lower solute content, once the elastic limit is exceeded, result in lower work hardening, lower tensile strengths, and the unusual increase in ductility. Of course, the electrical conductivity should, and does, increase.

EFFECT OF STRAIN AND SUBSEQUENT AGING ON THE ENDURANCE LIMIT

Since the supply of test stock of the original lots of 6, 10, and 20 Numbers Hard material had been exhausted, these fatigue tests were made on commercial 8 per cent tin-phosphorus-bronze, cold-rolled 8 Numbers Hard, in the "as-rolled" condition and after aging 100 hours at 250 degrees Cent. (435 degrees Fahr.). Since the specific aging temperature of 6 and 10 Numbers Hard material was 250 degrees Cent., it was assumed that material rolled to intermediate hardness should be aged at 250 degrees Cent. (435 degrees Fahr.).

The data are given in Fig. 4. The marked improvement of the aging treatment is clearly indicated. The practical value of the aging treatment is shown by the following interpretations:

1. If an engineering use requires a life of only 1,000,000 cycles of alternating stress, the as-rolled material will allow a maximum stress of 38,000 pounds per square inch as compared to 43,000 pounds per square inch for the aged material.

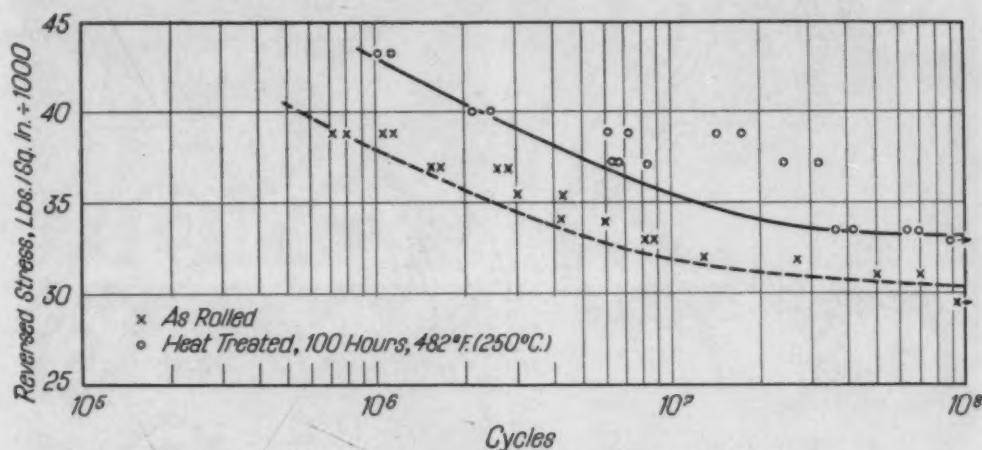


Fig. 4—S-N Curves. 8 Per Cent Tin Phosphor Bronze. 0.020 Inch Strip. 8 Numbers Hard.

2. If an application requires an alternating stress of 40,000 pounds per square inch, the as-rolled material will have a life of only 550,000 cycles as compared to 2,250,000 cycles for the aged material, a ratio of about 1 to 4.

3. If an engineering use requires a life of 100,000,000 cycles, the as-rolled material may be stressed as much as 30,000 pounds per square inch as compared to 33,000 pounds per square inch for the aged condition.

4. If an application requires an alternating stress of 33,000 pounds per square inch, the as-rolled material will have a life of 6 to 8,000,000 cycles as compared to 100,000,000 cycles for the aged material, a ratio of about 1 to 12.

Photomicrographs—The appended photomicrographs give evidence as to the grain sizes of the bronze in the cold-rolled and annealed conditions and show that no structural recrystallization has occurred following the aging treatments.

CONCLUSIONS

1. It seems well established that plastic strain induces a pre-

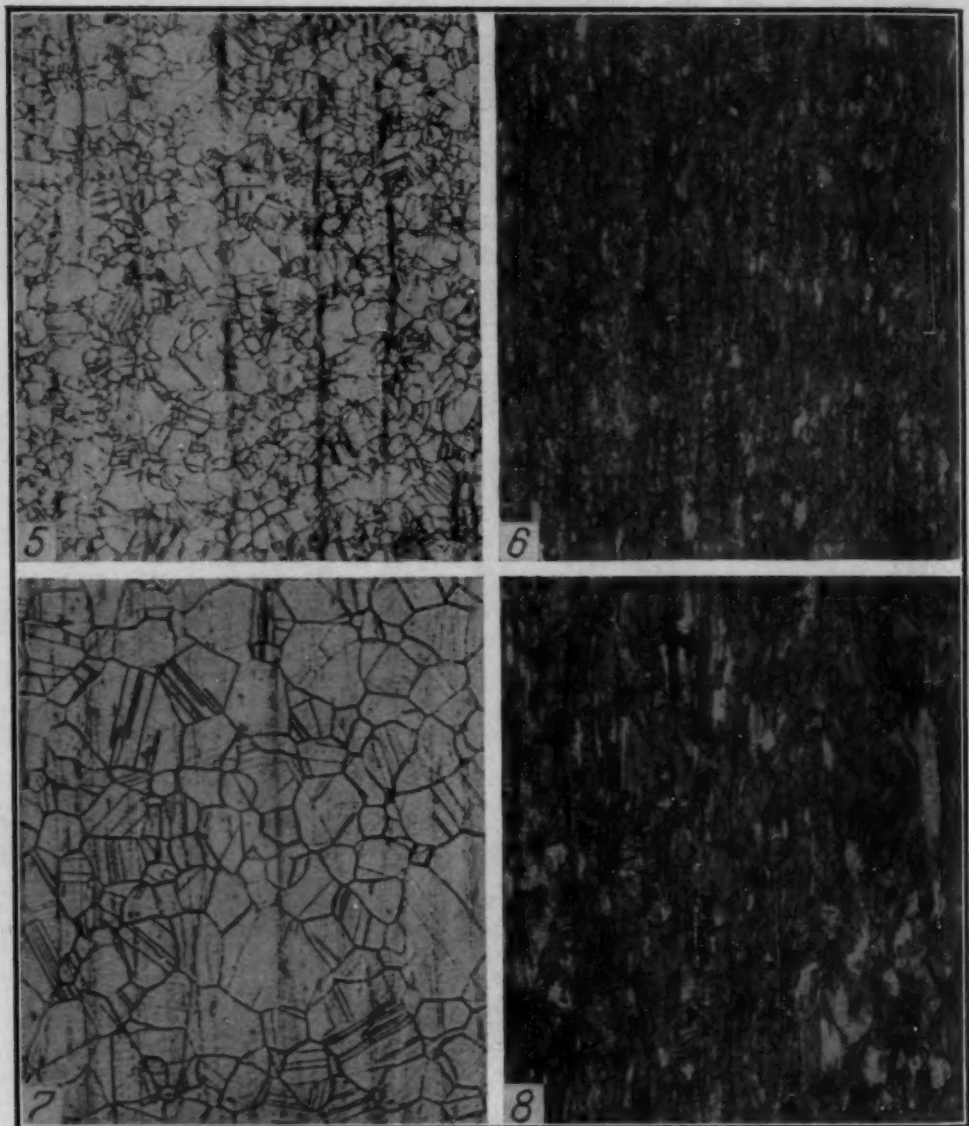


Fig. 5—6 Numbers. Fine Grain. Annealed Before Rolling. $\times 100$.
 Fig. 6—6 Numbers. Fine Grain. 4 Hours at 250 Degrees Cent. $\times 100$.
 Fig. 7—6 Numbers. Medium Grain. Annealed Before Rolling. $\times 100$.
 Fig. 8—6 Numbers. Medium Grain. 4 Hours at 250 Degrees Cent. $\times 100$.

cipitation reaction in bronzes at elevated temperatures below the recrystallization temperature range.

2. The induced precipitated phase acts as a blocking agent to the diffusion necessary for recrystallization. This results in higher service temperatures for the cold-rolled condition than would otherwise be obtained if the material were free to recrystallize unhindered at the lower temperatures.

3. Present data seem to indicate that this induced precipitation

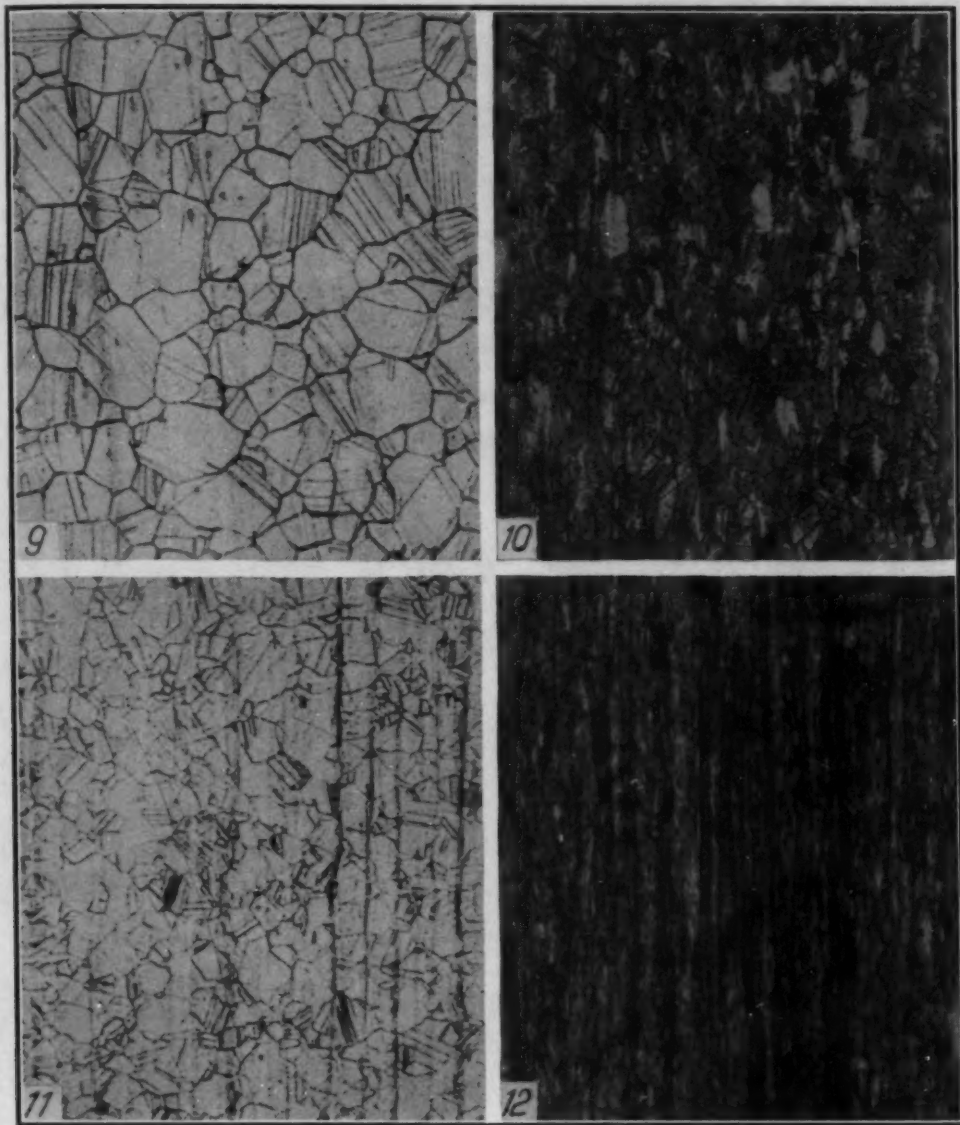


Fig. 9—6 Numbers. Coarse Grain. Annealed Before Rolling. $\times 100$.
Fig. 10—6 Numbers. Coarse Grain. 4 Hours at 250 Degrees Cent. $\times 100$.
Fig. 11—10 Numbers. Fine Grain. Annealed Before Rolling. $\times 100$.
Fig. 12—10 Numbers. Fine Grain. 4 Hours at 250 Degrees Cent. $\times 100$.

reaction should not be shown on the standard equilibrium diagram but should be plotted in a plane parallel to the standard equilibrium diagram, displaced from the standard along a third (a strain) axis, perpendicular to the junction of the temperature and composition axes.

4. Controlled use of the induced precipitation reaction yields marked improvement in the engineering properties of the cold-rolled solid solution bronzes. The proportional and elastic limits may be

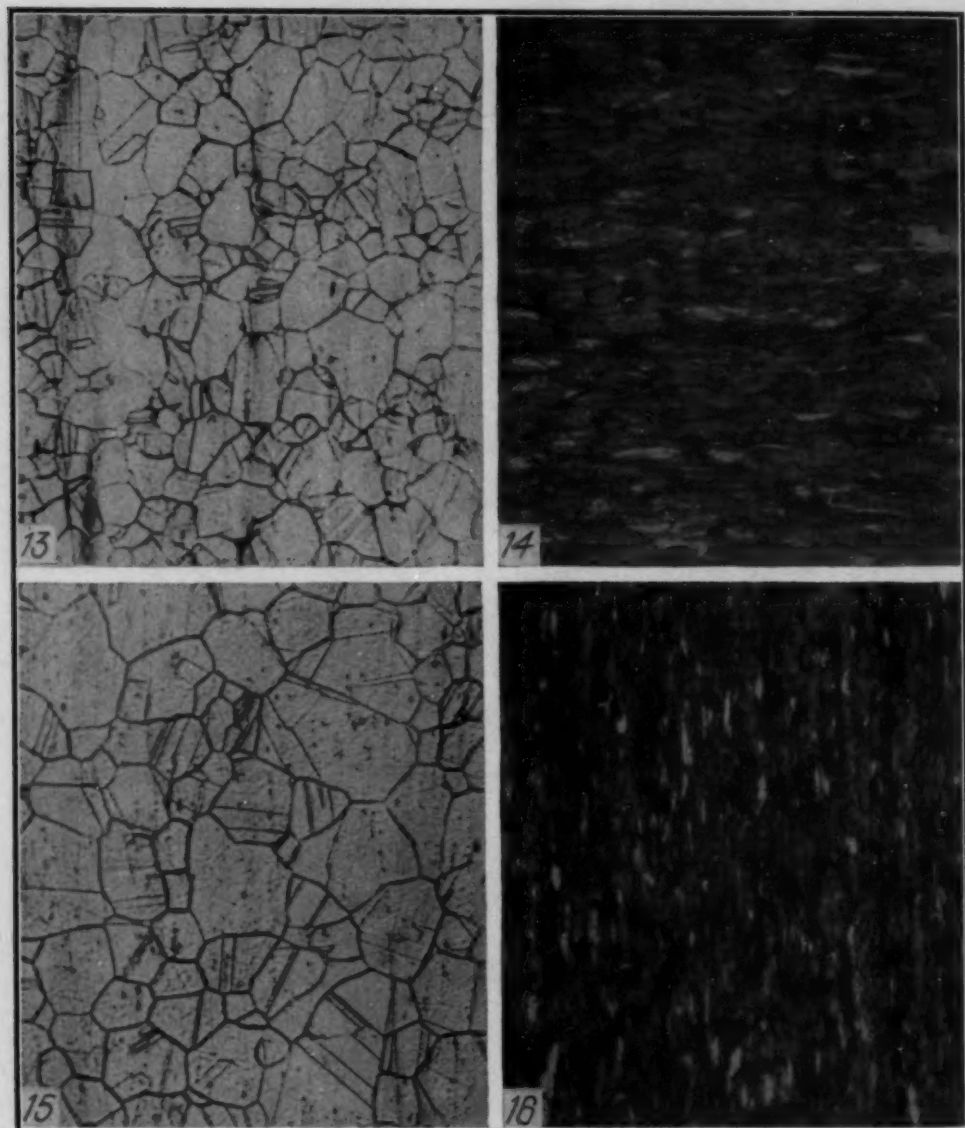


Fig. 13—10 Numbers. Medium Grain. Annealed Before Rolling. $\times 100$.
Fig. 14—10 Numbers. Medium Grain. 4 Hours at 260 Degrees Cent. $\times 100$.
Fig. 15—10 Numbers. Coarse Grain. Annealed Before Rolling. $\times 100$.
Fig. 16—10 Numbers. Coarse Grain. 4 Hours at 250 Degrees Cent. $\times 100$.

somewhat improved, or at least stabilized, in combination with a marked increase in elongation which should aid considerably in fabrication properties of the aged material. Resistance to fatigue is improved to a marked degree.

5. It is suggested that these effects result from strain-domain-precipitation within the grains of the cold-rolled material. This hypothesis requires further data and elaboration before it may become even a theory.

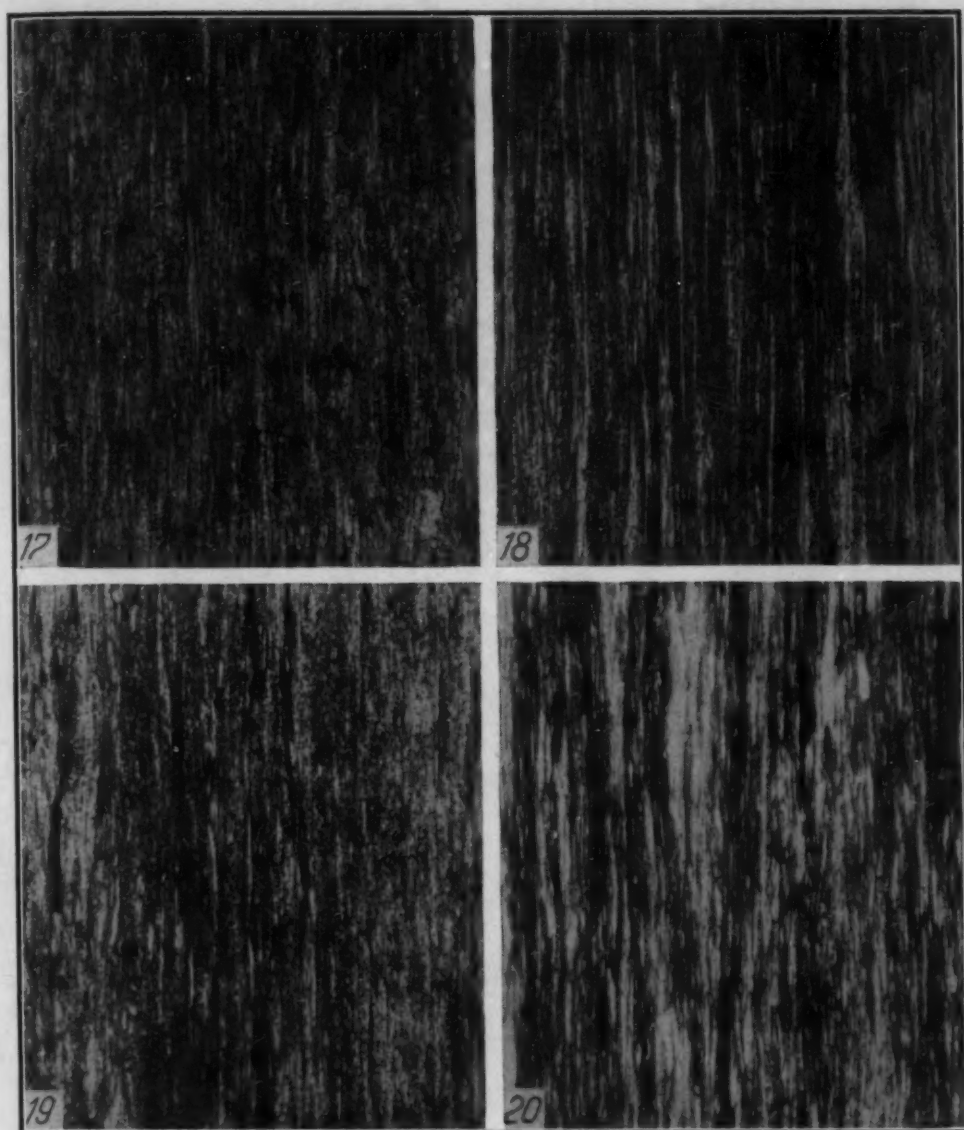


Fig. 17—20 Numbers. Hard. As-Rolled. $\times 100$.
Fig. 18—20 Numbers. Hard. 4 Hours at 225 Degrees Cent. $\times 100$.
Fig. 19—20 Numbers. Hard. 10 Hours at 225 Degrees Cent. $\times 100$.
Fig. 20—20 Numbers. Hard. 100 Hours at 225 Degrees Cent. $\times 100$.

6. Together with additional evidence (4), it is suggested that the principles and effects of this strain-induced-precipitation should apply in some degree to all strained solid-solution alloys.

ACKNOWLEDGMENTS

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Bibliography

1. Otto Dahl and Carl Haase, "Process for Increasing the Electrical Conductivity of Tin Bronzes," U. S. Patent 2,128,122.
2. Owen and Iball, "X-Ray Investigation of Certain Cu-Sn Alloys," *Journal, Institute of Metals*, Vol. 57, 1935, p. 267, and Vol. 58, 1936, p. 283.
3. C. Haase and F. Pawlek, "Zur Kenntnis der Kupfer-Zinnlegierungen," *Z. Metallkunde* Vol. 28, 1936, p. 73.
4. R. H. Harrington, "The Role of Strain in Precipitation Reactions in Alloys," *Age Hardening of Metals*, published by American Society for Metals, 1940.

HARDENING CHARACTERISTICS OF VARIOUS SHAPES

BY M. ASIMOW AND M. A. GROSSMANN

Abstract

Prior studies of round bars had suggested a method for judging severity of quench as well as a criterion of hardenability. These principles are now applied to shapes other than rounds. For flat plates a set of charts is developed, which is entirely analogous to those previously calculated for rounds, and is found to accord closely with experimental data. For irregular shapes not subject to precise calculation, an approximation is suggested, in which each point in the irregular shape is considered to behave like a corresponding point in a round bar of "equivalent diameter."

HARDENING CHARACTERISTICS OF VARIOUS SHAPES

WHEN judging the hardening behavior of a steel from hardness traverse or hardness distribution curves on round bars as illustrated in Fig. 1, it is recognized that the extent of hardening is affected both by the hardenability of the steel and by the severity of the quench. It has been shown in the case of round bars (1), (2)¹ that it is possible to deduce from the curves the severity of quench that was employed, thus making it possible to judge the true hardenability of the steel, referred to an ideal quench. This further makes it possible, once the severity of a quench is known, and when the hardness distribution in one size is known, to predict within certain limits the hardness distribution which would be obtained in some other size, or with some other severity of quench.

This consideration of round bars, though thus very useful in judging the hardenability of a steel and the severity of a quench, has limitations in any quantitative application to commercial shapes, since but few such parts have the shape of simple circular cylinders. It is the purpose of the present paper to extend this treatment to shapes other than round bars. In the case of plates, a system is developed

¹The figures appearing in parentheses pertain to the references appended to this paper.

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which is entirely analogous to the one for round bars and which has the same degree of accuracy. For irregular shapes, a method of analysis is suggested which provides close approximations.

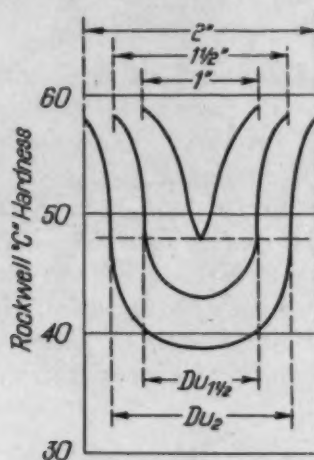


Fig. 1—Hardness Distributions in 3 Sizes of Rounds, from Which Severity of Quench Can Be Estimated.

By way of necessary review, it may be pointed out that in the scheme for round bars recognition of the following points constitutes the essence of the system:—

1. As already mentioned, it is well understood that hardenability and severity of quench are the factors which determine the depth of hardening. These factors must be determined separately. Thus suppose that some laboratory reported that a 2-inch round had been hardened, and that an unhardened core was found whose diameter was $\frac{9}{10}$ of the diameter of the bar (depth of hardening was 0.1 inch), and

Table I

Diam. of Bar, Inches	Relation of Unhardened Diameter to Bar Diameter, Per Cent	
	Quenching Severity $H = 1.0$	$H = 5.0$
2.00	90%	90%
1.75	87%	88%
1.50	80%	86%
1.25	65%	82%
1.00	25%	76%

that another laboratory reported that some other 2-inch bar had been hardened, finding the same extent of unhardened core. Curiosity would be aroused as to whether the steel had the same hardenability with the same quench, or different hardenability and therefore

different quenches. To determine this, each laboratory would harden a series of sizes of its steel in its quench. The results might come out as in Table I.

The two sets of points are plotted as the round and square dots of Fig. 2, and by comparing them with the standard chart (1), (2) it is found that in the first case the quenching severity is about

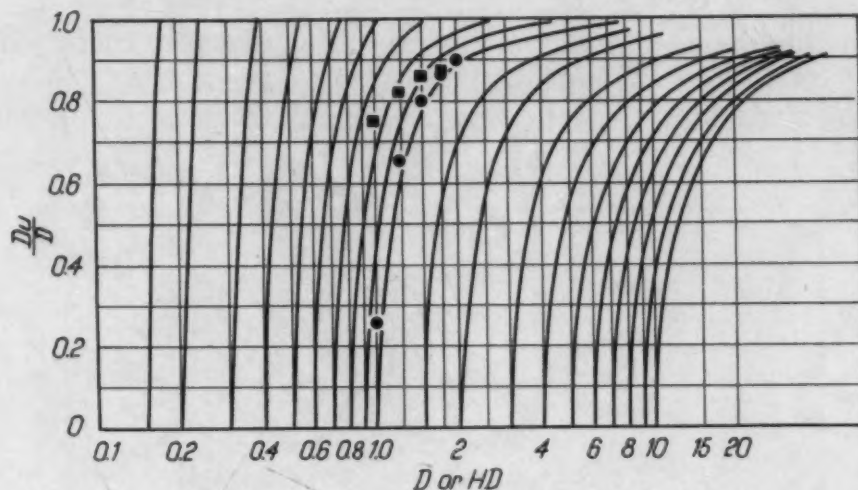


Fig. 2—Curves for Estimating Severity of Quench in Rounds.

$H = 1.0$ (a very mild water quench or a strong oil quench) and in the second case the quenching severity is about $H = 5.0$ (a rather strong water quench). The respective hardenabilities would be found to be $D_I = 1.70$ inch and 0.75 inch (see next, item 2).

2. An accurate measure of hardenability is the "critical size," which is the bar diameter which is just "hardened throughout" as seen in an etch test (largest bar which shows no unhardened core; it is just half-hardened (50 per cent martensite) at the center of the bar). Having determined this critical size in any actual quench, and having ascertained the severity of the quench, the "ideal critical size D_I " is read from a diagram such as Fig. 3. This value D_I is the critical size which would be obtained in the severest possible quench, and is thus a measure of hardenability alone. The purpose in using it is to speak a common language for hardenability for different laboratories or different quenching techniques. All critical sizes obtained in any actual test of a steel will of course be smaller than this value D_I , since D_I represents the greatest degree of hardening of which that steel is capable, as brought out by a theoretically perfect quench.

3. By the application of these principles it is possible to predict hardness distributions which would be obtained in other bar sizes or in other severities of quench, provided only that the hardness distribution is known for one bar size for which the severity of quench is also known (1).

THE BEHAVIOR IN PLATES

In considering the heating and cooling of flat plates and round bars, T. F. Russell (3) published tables showing calculated time-temperature distributions, making the same assumptions as those which

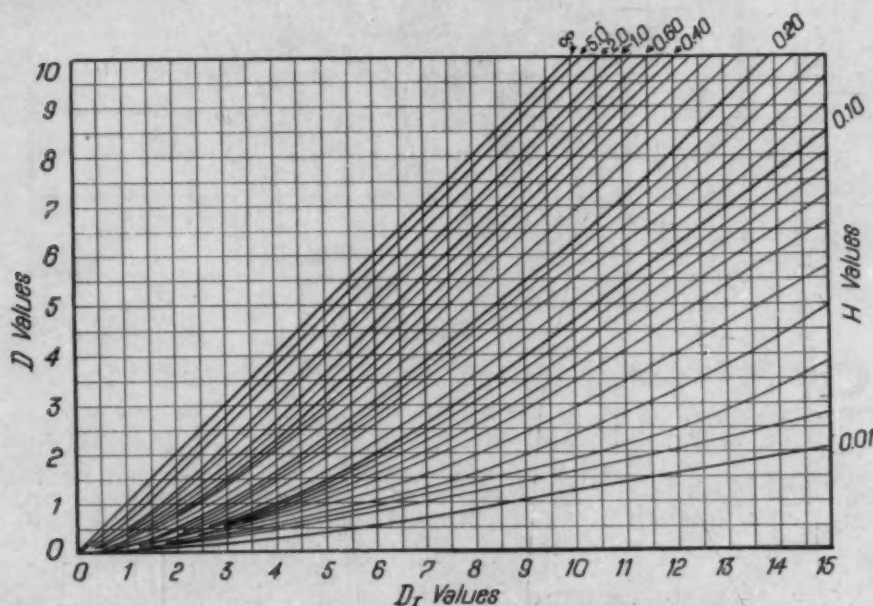


Fig. 3—Relationships Among Ideal Critical Size, D_i , Actual Critical Size, D , and Severity of Quench, H .

had been made in the present work and which resulted in the curves of Figs. 2 and 3 herewith, namely, constant heat transfer factor (Newton's law of heating or cooling) and constant diffusivity in the steel. It is therefore possible to set up an entirely analogous system for flat plates by using Russell's figures for plates, interpolating to find the values for "half-temperature time." This latter value, which is the time to cool from the quenching temperature to a temperature half-way down to that of the quenching liquid, was adopted as a criterion of hardening because it accorded closely with the test data, and because it has a reasonably good theoretical background (1).

The use of Russell's calculations for plates, then, in the present system, results in Figs. 4, 5 and 6, where H is the quenching severity

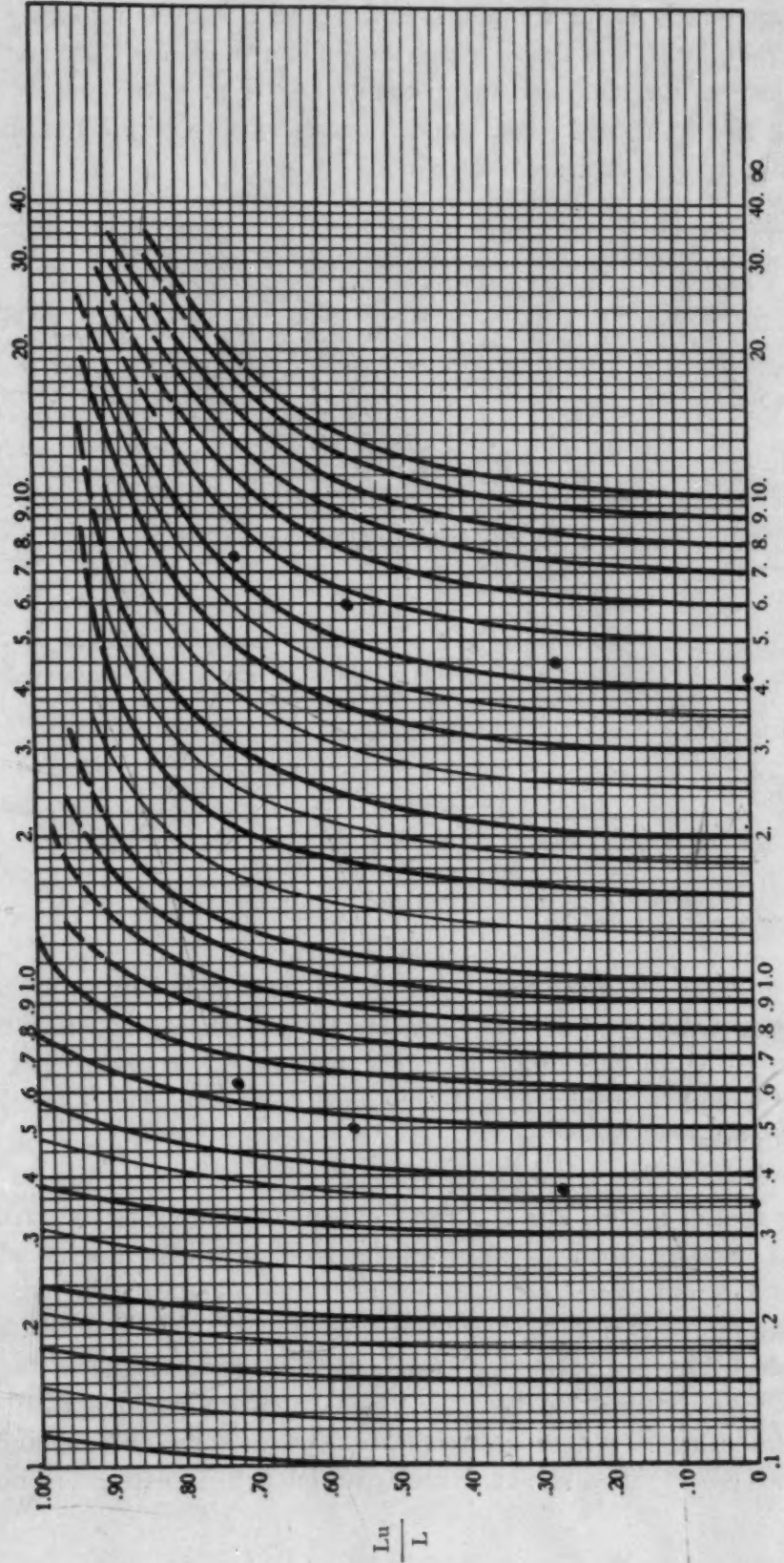


Fig. 4—Curves for Estimating Severity of Quench in Plates.

in the same units as for rounds, L is the plate thickness replacing bar diameter D , and L_I is the "ideal critical plate thickness" used in the same sense as the ideal critical diameter D_I . Fig. 4, of course, corresponds to Fig. 2, and Figs. 5 and 6 correspond to Fig. 3 but showing two ranges of values.

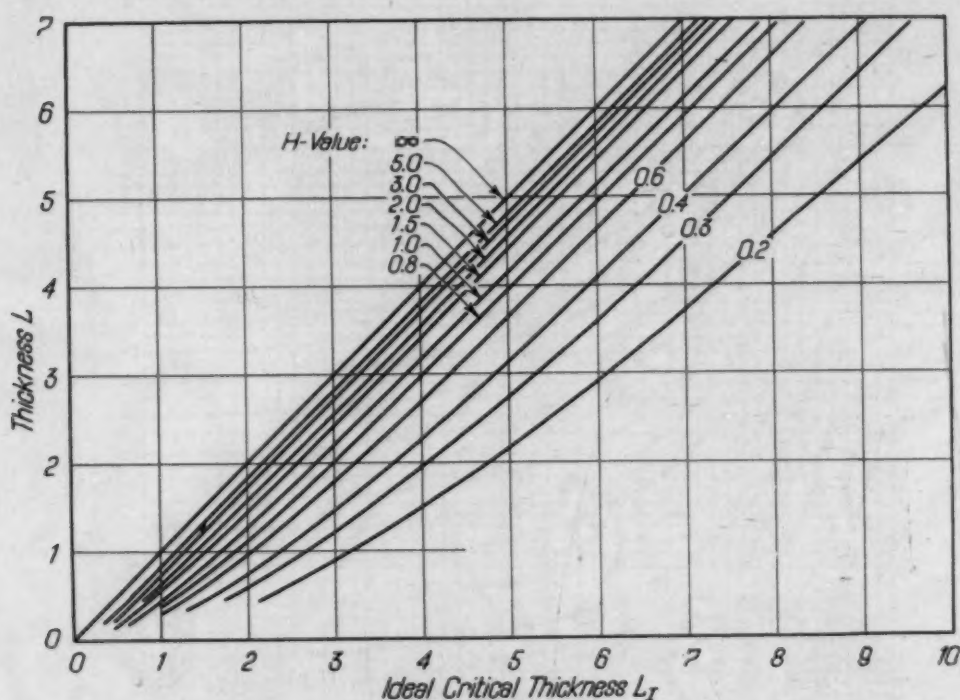


Fig. 5—Relationships Among Ideal Critical Plate Thickness L_I , Actual Critical Plate Thickness L and Severity of Quench H .

In addition to these values, it would be convenient to be able to translate from ideal critical diameter D_I to ideal critical thickness L_I , so as to be able to judge from the behavior in rounds what the behavior would be in plates, and vice versa. The following relationship is found:

$$D_I = 1.377 L_I$$

meaning that if a steel has a hardenability such that its ideal critical thickness in plate form is 1.00 inch, its ideal critical diameter when tested as a round would be 1.377 inch. For convenience in use, the relationship is shown as the straight line in Fig. 7. (Caution: this relationship does not apply for other severities of quench).

It was of interest to compare these calculated values with experimental values. A test case was provided in some data published by Luerssen (4), who shows a series of quenched plates, as photo-

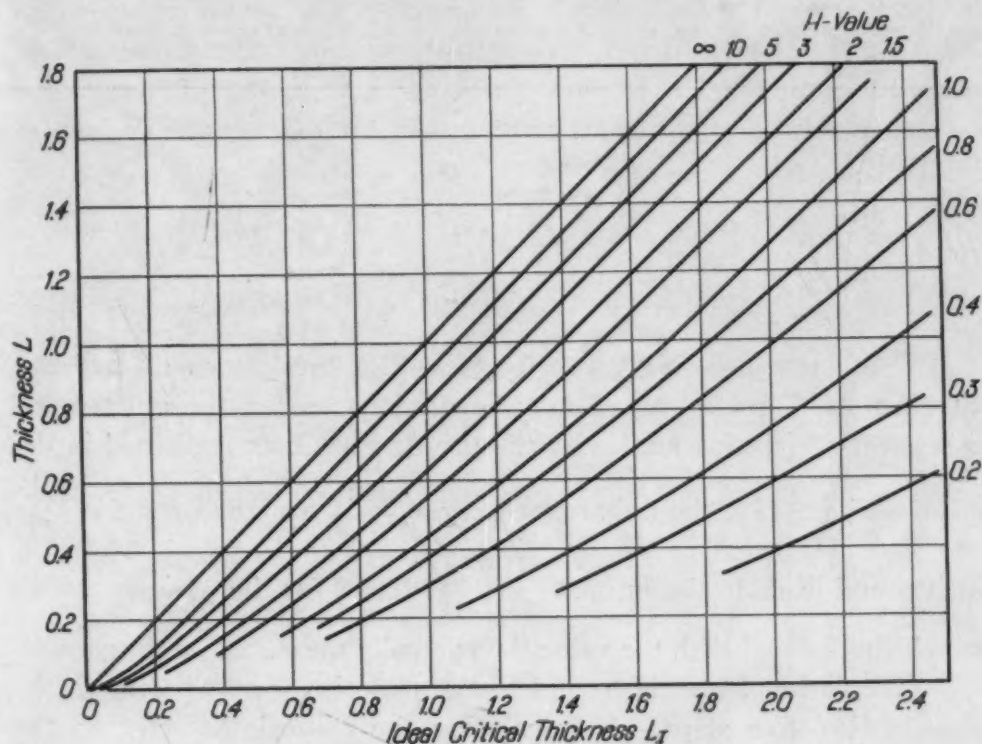


Fig. 6—Same as Fig. 5 but More Open Scale.

graphs of etched sections showing unhardened cores (his Fig. 2), and a variety of sections from a single steel, including two sizes of rounds and one plate (his Fig. 3). The severity of quench can be estimated from the unhardened cores of the plates, following values having been measured on the photographs:

Table II

Stated Sizes of Thickness of Plate, L, Inch	Proportion of Unhardened Core
	Measured on Photograph, $\frac{L_u}{L}$
11/32 = 0.344	0
12/32 = 0.375	0.27
16/32 = 0.500	0.56
20/32 = 0.625	0.72

From these values, using Fig. 4 herewith, it is found that the severity of quench was about $H=12$, as indicated by the dots in Fig. 4.

Considering now his etched sections of two rounds and one plate of identical steel, the following values were measured on the photograph:

Table III

1¼-inch round,	$\frac{Du}{D} = 0.825$
¾-inch round,	$\frac{Du}{D} = 0.68$
⅝-inch plate,	$\frac{Lu}{L} = 0.73$

Using the quenching severity $H = 12$, the known unhardened core on the 1¼-inch round may be used to predict the unhardened core on the ¾-inch round. By the use of Fig. 2, as explained in detail elsewhere (2), the unhardened core $\frac{Du}{D} = 0.825$ for the 1¼-

inch round leads to the estimate $\frac{Du}{D} = 0.665$ for the ¾-inch round, which checks well with the value 0.68 actually measured on the sample.

Calculating from the round to the plate involves what might be classified as five steps,—1. calculating the critical diameter in the actual quench; 2. from this, the critical diameter in the ideal quench; 3. from this the corresponding critical plate thickness in the ideal quench; 4. then the critical plate thickness in the actual quench; and finally 5. the unhardened core is the known plate thickness. The steps are carried out very easily on the charts of Figs. 2 to 7.

1. The critical diameter in the actual quench is found by the use of Fig. 2; if the 1¼-inch bar has an unhardened core $\frac{Du}{D} = 0.825$, and $H = 12$, the critical diameter, where the unhardened core is just absent, will be $D_0 = 0.500$ inch.

2. The critical diameter in the ideal quench is found by the use of Fig. 3, for if $D = 0.500$ and $H = 12$, then the ideal diameter is read on the diagram as $D_I = 0.580$ inch.

3. The plate thickness in the ideal quench is then read from Fig. 7; corresponding to $D_I = 0.580$ we find the ideal plate thickness is $L_I = 0.421$ inch.

4. The critical plate thickness in the actual quench is then read from Fig. 6; at the point where $L_I = 0.421$ and the quench is $H = 12$, we find the plate thickness $L = 0.355$ inch.

5. Finally the unhardened core in the actual ⅝-inch plate is estimated from Fig. 4. Since $H = 12$ and the critical thickness

$L_0 = 0.355$ (see above), $H \times L_0 = 4.26$. Follow the curve which would rise from the origin at 4.26 (between the curves 4.0 and 5.0 at the origin) until it intersects the vertical straight line 7.50; this value 7.50 represents the actual plate, since thickness $L = \frac{5}{8}$ -inch = 0.625, and $H = 12$, so that $H \times L = 12 \times 0.625 = 7.50$. The intersection shows that the value $\frac{Lu}{L}$ is thus estimated to be 0.72, which

agrees very closely indeed with the value 0.73 read from the photograph (see Table III).

It has thus been possible, through knowing the depth of hard-

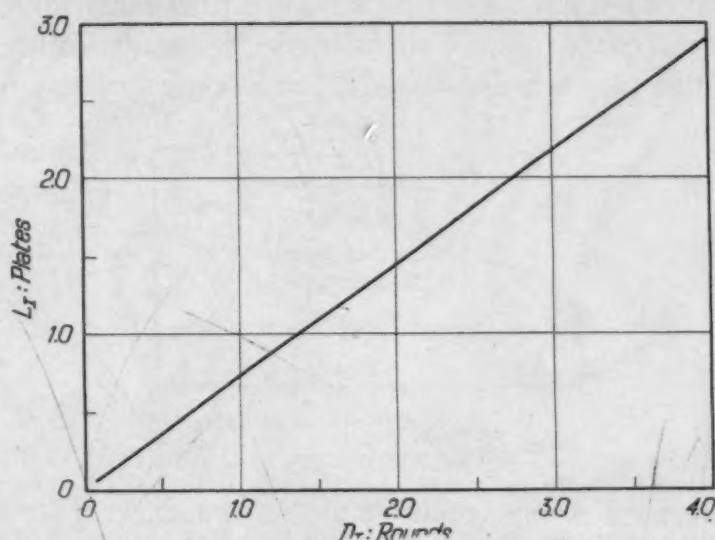


Fig. 7—Relation Between Ideal Bar Diameter D_1 and Ideal Plate Thickness L_1 .

ening on a round bar and knowing the severity of quench, to predict quite accurately the depth of hardening when that steel is used in the form of a plate. Such agreement with the theory, which has been checked also on other data converting from rounds to plates, indicates that Figs. 2 to 7 may be used with considerable confidence.

For convenience in practical testing, furthermore, this points to the feasibility of using simple billet slices for hardenability tests, avoiding the necessity for further machining of test pieces. Such slices would be thinner than rounds for the same data, and by this system the derived data could be converted to the values for rounds.

THE PROBLEM OF IRREGULAR SHAPES

It is now proposed to apply similar principles to the treatment

of irregular shapes. When the hardness distribution in some irregular shape has been determined for one steel and one quench, it would be convenient to be able to predict results for that shape when using a steel of some other hardenability or when employing some other severity of quench. The problem may be considered under three headings,—

A—Different hardenabilities (quench unchanged)

B—Different quenches (steel unchanged)

C—Different hardenability and different quench.

A. *Different Hardenabilities*—This is an extremely simple case, involving merely the substitution of different hardness values for the values already found. If we consider the important feature that the quench in this case remains unaltered, it is clear that any individual

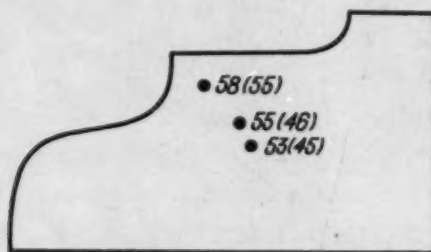


Fig. 8—Hardnesses in an Irregular Shape.

point in the piece always cools in the same manner (same cooling time),—it is only the hardenability that is changed. Any simple hardenability test may therefore be run and the hardness in the new steel substituted for the old. For example, consider the irregular shape of Fig. 8, which at the three indicated points was found to have the hardness 58, 55 and 53. Now suppose an ordinary hardenability test is run by quenching two rounds, one of them marked 1, in Fig. 9, being the steel of Fig. 8, and the other being the new steel proposed to be used, marked 2 in Fig. 9. The point on curve 1 which has a Rockwell hardness $R_c = 58$ in the test of Fig. 9 must have had the same cooling time as the point marked 58 in Fig. 8, since this cooling time resulted in the same hardness $R_c = 58$ in the two cases, the steel being the same. Now according to Fig. 9, that cooling time (same position in the bar) when using the new steel resulted in a hardness of $R_c = 55$. It is not necessary that the quench of Fig. 9 be the same as the quench of Fig. 8. It is however necessary that the bars (1) and (2) of Fig. 9 be quenched alike. It is clear then

that the position in Fig. 8 which had a hardness of 58 would, when using the new steel, show a hardness of $R_c = 55$. In the same way, points 55 and 53 would show hardness of 46 and 45 respectively, when using the new steel, as shown by Fig. 9 and indicated in parentheses in Fig. 8.

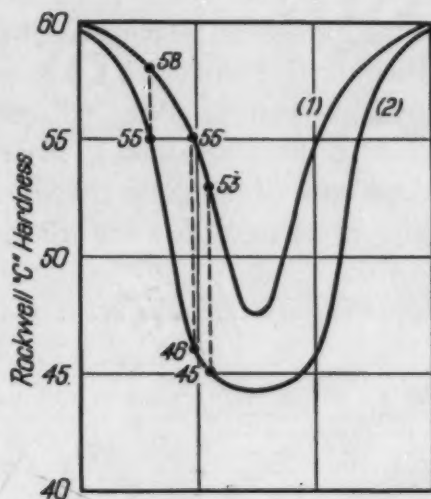


Fig. 9—Hardness Distributions in Two Steels, for Substitution in Fig. 8.

In the same way, a complete hardness survey of the original piece would provide a complete plan for the new hardness values, and this will be illustrated in greater detail in the case of rectangular prisms.

B. Different Quenches—For an irregular shape such as Fig. 8, a change in *quench* is a decidedly more complex problem, and must be discussed in considerable detail. The solution suggested here is based on the following concept: In considering Figs. 8 and 9, it was pointed out that the hardness at point 58 in Fig. 8 was the same as that at some position in the round of Fig. 9. In that case, where the quench was unchanged, the position in the irregular shape behaved the same as some position in the round. In the new case, where the quenching severity is changed, a similar principle is employed, in that a position in the irregular shape is considered to behave the same as a corresponding position in a round, but now the size of round becomes important. That this is true becomes apparent from a study of Fig. 2, which shows that the effect of the quenching severity changes with the size of the round. Hence, if one desires to estimate the effect of severity of quench for a particular point in an irregular

shape, it is necessary first to ascertain the size of round which can be considered equivalent, at the original quench. Thereupon, since the behavior in rounds follows known laws (Figs. 2 and 3), it is easy to estimate quite closely the probable effect of a change in quench. One has thus arrived at the concept of "equivalent diameter," in which each point in an irregular shape is considered to behave the same as the corresponding point in an equivalent diameter of round.

The problem is to find for each point the diameter which gives equivalent behavior. In the case of plates (discussed above), and of rectangular prisms (to be discussed later), the relationships can be calculated exactly. In the case of irregular shapes not subject to calculation, the relationships are based on a knowledge of the hardening characteristics of the steel.

C. Different Hardenability and Different Quench—This case is

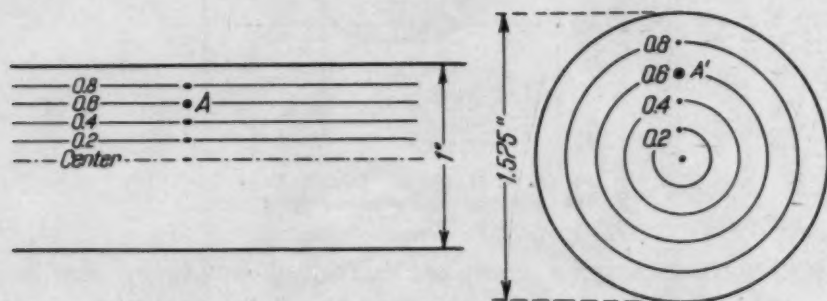


Fig. 10—Equivalent Diameter for Point A in Plate.

merely a combination of cases A and B, and the solution is provided by calculating for the new quench and then substituting for the new hardenability.

EQUIVALENT DIAMETERS FOR PLATES

The method of equivalent diameters can be developed for plates as for any other shape. As shown above (Figs. 4 to 7) it is not necessary in practice to use this indirect method, since the calculations can be made directly, but it will be useful for the subsequent development of the method to examine it when thus applied to plates.

In Fig. 10, representing a cross section of a plate 1 inch thick, attention may be directed for example to point A, which is at a position 0.6 of the way from the center to the surface. As indicated above, point A may (within certain limits) be considered to behave like a corresponding point A', which is similarly situated at a position 0.6 of the way from the center to the surface in a round bar, Fig. 10.

The problem is to find the size of bar which has the equivalent behavior at the 0.6 point. Equivalent behavior means, when using a specific quench, that the 0.6 position in the plate would have the same hardness as the 0.6 position in the equivalent round (assuming the same steel).

Since the relationships between plates and rounds have been worked out completely, the size of round can be calculated by the use of Figs. 2 to 7. It is furthermore possible in this particular case to estimate the reliability when using this round for the calculation of other quenches.

The calculations may be followed step by step in Table IV.

Table IV

Plate 1 inch thick, thickness $L = 1.0$.
Severity of quench $H = 1.6$.
Then $HL = 1.6 \times 1.0 = 1.6$.
Read on Fig. 4 from the 0.6 position on the vertical line $HL = 1.6$ down along the projected curve to the origin, to find the point having the same behavior at the center of the plate. The point is found to be $HL_0 = 1.32$.
Then $L_0 = \frac{HL_0}{H} = \frac{1.32}{1.6} = 0.825$ inch.
This is the plate thickness of which the center would have the same cooling-time (and in an actual test would therefore show the same hardness) as the 0.6 position in a 1-inch plate, when the quench is $H = 1.6$.
For an infinite quench, use Fig. 5. At $L_0 = 0.825$ and $H = 1.6$, there is found $L_1 = 1.29$ inch, which is the plate thickness which would again have this same hardness at its center after an "ideal" quench (severest possible).
For the corresponding round, read from Fig. 7, which shows the ideal critical diameter corresponding to the above value $L_1 = 1.29$, the new value being $D_1 = 1.79$. This is the diameter of round which in an ideal quench would again have the same hardness at the center.
For the round in the actual quench $H = 1.6$, use Fig. 3. $D_1 = 1.79$ (see above), therefore read from Fig. 3 at $D_1 = 1.79$ and $H = 1.6$, $D_0 = 1.23$, this being the size round whose center again has the same hardness when quenched $H = 1.6$.
<i>Equivalent diameter.</i> It is now possible to find the round whose 0.6 position will have the same hardness. To use Fig. 2, we have $D_0 = 1.23$, and $H = 1.6$. Therefore $HD = 1.6 \times 1.23 = 1.97$. Follow the curve from the base-line position 1.97, upward to the right until it reaches the horizontal $\frac{Du}{D} = 0.6$,
which occurs at the position where the vertical line is 2.52.
This is the HD value for the desired round, so that the
diameter $D = \frac{HD}{H} = \frac{2.52}{1.6} = 1.575$. The equivalent diameter is therefore 1.575 inches,
this being the size round whose 0.6 position has the same hardness as the 0.6 position in the plate, when both are quenched $H = 1.6$.

Table IV states that the 0.6 position in a 1-inch plate acquires the same hardness as the 0.6 position in a 1.575-inch round (using the same steel) when the quench is $H = 1.6$. When the quench is changed, the 1.575-inch round may still be used to estimate the behavior in the plate, as indicated in the calculations of Table V and Fig. 11. The purpose is to compare the true changes (calculated directly on the plate) with the estimated changes (calculated from the equivalent round). Fig. 11 shows that, when the quench does not

depart too far from the original value $H = 1.6$, the concordance is very close indeed, as for example in the range $H = 1.0$ to $H = 3.0$, where the use of equivalent diameter provides a very fair approximation. Since in commercial applications for a particular part the variation in quenching severity is not very great, the variation at the extremes of Fig. 11 is not of practical importance.

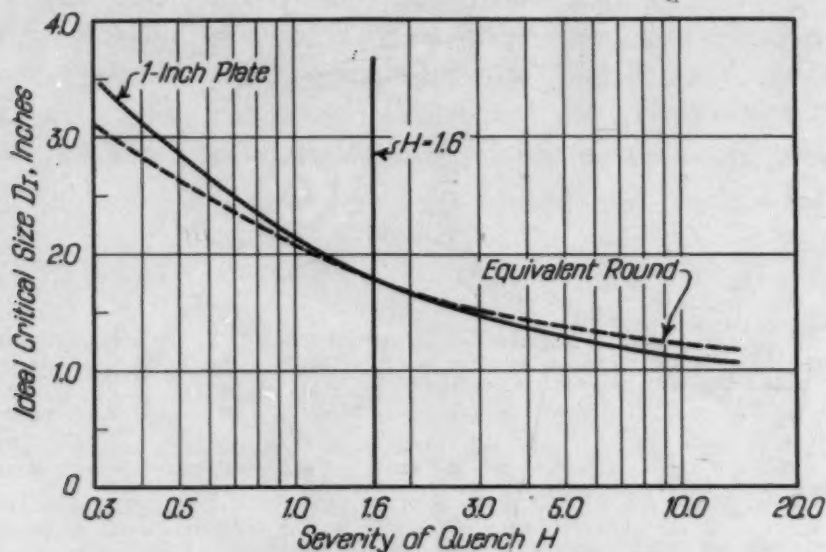


Fig. 11—Indication of Accuracy, When Using Equivalent Diameter for Judging Behavior of Plate.

In Fig. 11, the ordinate ideal diameter D_r should be considered as being related to its center cooling-time, and consequently related to its center hardness. The variation could be stated just as well in terms of hardness for a particular steel, assuming that the relation between hardness and ideal diameter (or time) is known for that steel.

A point not to be overlooked is that the equivalent diameter for one position (such as the 0.6 position in Fig. 10) will not be valid for some other position. The equivalent diameters for other positions can be calculated in the same way as before, and give the following values for a 1-inch plate quenched $H = 1.6$:

Position	Equivalent Diameter D_r , Inches
$\frac{L_u}{L}$	
0.2	1.47
0.4	1.53
0.6	1.575
0.8	1.58

Table V

1-Inch Plate			Equivalent Round									
H	HL	HL _o	At 0.6 Position, Point Behavior Like Center of		Con-stant Quenches Equiv- alent Diam- eter		Same as for Plates		From 0.6 Position,			
			H _o =		D _i =		H		HD _o =			
			L _o =	L _i	D _i =	D	H	HD	D _o	D _i		
14.	14.	10.0	0.714	0.77	1.06	1.575	14.	22.1	15.3	1.09	1.16	
8.2	8.20	6.0	0.732	0.845	1.16	1.575	8.2	12.9	9.2	1.12	1.26	
5.3	5.30	4.0	0.755	0.92	1.27	1.575	5.3	8.35	6.2	1.17	1.36	
2.46	2.46	2.0	0.813	1.14	1.57	1.575	2.46	3.87	3.0	1.22	1.58	
1.6	1.6	1.32	0.825	1.29	1.79	1.575	1.6	2.52	1.97	1.23	1.79	
0.91	0.91	0.80	0.88	1.62	2.23	1.575	0.91	1.43	1.18	1.30	2.15	
0.54	0.54	0.50	0.926	2.00	2.76	1.575	0.54	0.85	0.73	1.35	2.54	
0.315	0.315	0.30	0.953	2.50	3.44	1.575	0.315	0.50	0.45	1.43	3.05	

RECTANGULAR BLOCKS

In the light of the above argument as applied to plates, it is useful to apply the same system to rectangular blocks. This is a peculiarly instructive case because heat flow in such a piece during quenching follows a more complicated pattern, and so provides a scheme which may be applied to still more complicated shapes. That is to say, in quenching flat plates or round bars, the heat flow takes place along straight lines perpendicular to the surface of the piece, whereas in rectangular blocks and in irregular shapes, the heat flow is along curved lines. This case is also useful because it is still subject to mathematical analysis, thus providing a check, a feature which is no longer true in very complicated shapes, such as for example a crankshaft.

COMPARISON OF BLOCKS AND BARS

It was pointed out above, in setting up the system for flat plates and its comparison with round bars, that this was based on equal cooling times (actually "half-temperature cooling times") in the plates and the bars. That is to say, if a point in a plate had a certain cooling time, resulting in a certain Rockwell hardness, then a point in a bar having the same cooling time would have the same hardness, and it was shown that the system was reliable when checked with experimental data. This portion of the plate study was therefore a direct comparison of the effects of heat flow in plates and round bars. The other portion of the study involved the concept of "equivalent diameter," which showed how calculations for a plate can be made by studying the behavior of a bar of appropriate size.

The same scheme is followed here for rectangular blocks,—first a direct comparison of blocks and bars, and second an estimate of the behavior of blocks judged solely by the behavior of bars of equivalent diameter.

DIRECT COMPARISON

The somewhat complicated case of heat flow in rectangular blocks is based on the principles set forth in the Appendix. By means of such analysis, the cooling times may be calculated for the different positions in a block.

Before proceeding to the effect of severity of quench, it may be well to refer again to Figs. 8 and 9, where it was pointed out that a mere change in hardenability, with the quench unchanged, is an extremely simple case. It is merely necessary to substitute, at each point in the hardness survey, the new hardness obtained with that particular cooling time.

Table VI
Compositions of Steels

	C	Mn	P	S	Si	Ni	Cr	Mo
1045	0.47	0.86	0.012	0.050	0.23
1090	0.97	0.36
T-1340	0.39	1.74	0.023	0.021	0.26	0.01	0.13
Amola	0.69	0.81	0.014	0.024	0.24	0.01	0.00	0.22

In a rectangular block, the hardness survey could be plotted as lines of equal hardness, like a topographic map. When changing to a steel of different hardenability, each line would retain its same shape and would merely represent a different hardness.

This was illustrated in the present case by quenching four steels, whose compositions are given in Table VI. They were treated as rectangular blocks, $2\frac{1}{2}$ inches by 4 inches by 6 inches, with a severity of quench $H = 1.6$. The results of the hardness surveys are shown in Fig. 12, where the hardnesses for each block are shown at the same positions. In accordance with the above discussion, it is seen that in going from T-1340 for example to S.A.E. 1090 steel, it would be permissible and reliable to use the hardness distribution lines as found in the T-1340 steel, merely assigning the new hardness numbers as found for the S.A.E. 1090 steel. The hardness lines were drawn with the same shape, and it is seen that the experimental points fall on these lines with a good degree of precision.

The next step is to compare the hardenability of a steel as deduced from quenched blocks with the hardenability as judged from quenched rounds of the same steel. The analysis described in the Appendix makes possible the calculation of half-temperature cooling

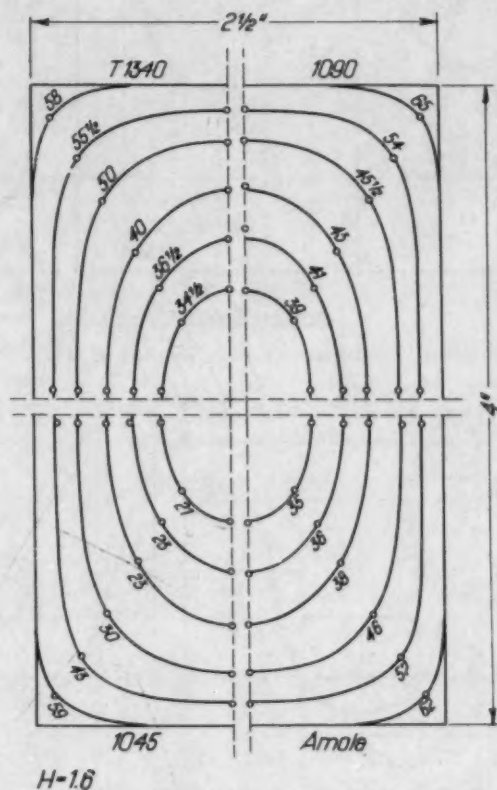


Fig. 12—Hardness Distributions in Blocks, Using Four Steels, Severity of Quench $H = 1.6$.

times at different positions in the blocks, and for the round the method has already been described (1). Such a comparison makes it possible to compare the blocks with the rounds, since a particular cooling time at a certain position in the block should give the same hardness as that same calculated cooling time at its position in the round. The hardness distributions for the blocks were shown in Fig. 12, and those for the rounds are shown in Figs. 13 and 14.

Actually, the cooling times are found in terms of "ideal diameter D_i " in rounds, which may be translated to cooling times, in seconds, by assigning a value for the diffusivity. From calculations provided by Austin (5), and after comparison with some experimental data, a figure 0.009 square inch per second was adopted as a working

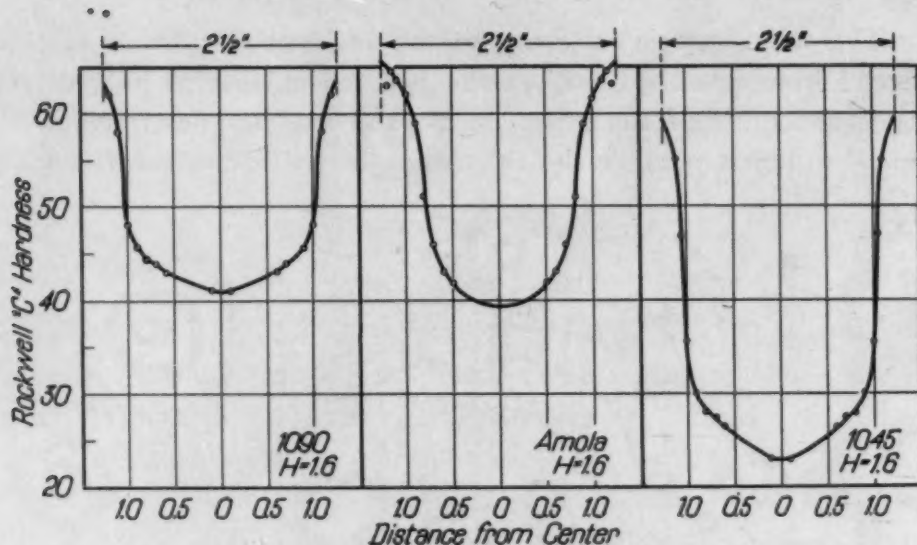


Fig. 13—Hardness Distributions in $2\frac{1}{2}$ -Inch Rounds of Three Steels, $H = 1.6$.

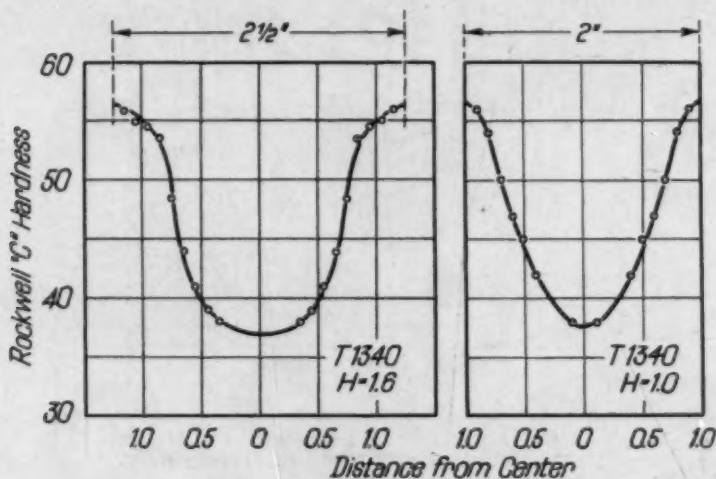


Fig. 14—Hardness Distributions, One Steel, $2\frac{1}{2}$ -Inch Round with $H = 1.6$, 2-Inch Round with $H = 1.0$.

value for the diffusivity, while realizing that it is not constant, and is different for austenite, pearlite and martensite. Using this value, Fig. 15 may be used for reading the half-temperature time from the values of "ideal diameter D_I ".

As stated above, it is desired to compare the "hardenabilities" of the two steels as judged from blocks and from rounds. But the hardenability of a steel may be defined as the hardnesses produced in that steel by various cooling times when quenched. In other words, hardenability may be presented as a graph of hardnesses vs. cooling times. Therefore, if the above calculated cooling times are plotted against the actual hardnesses found in the quenched blocks and

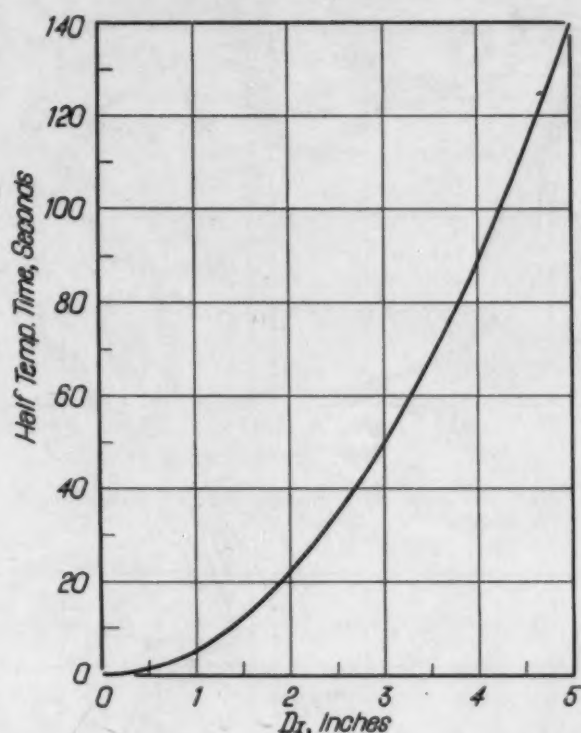


Fig. 15—Relation Between Time of Center-Cooling and Diameter of Round, Ideal Quench, Diffusivity 0.009 Sq. In. Per Second.

rounds, we have a comparison of the hardenabilities as judged from the two different quenched pieces. They should agree.

The results for blocks and rounds for the four steels are as shown in Figs. 16 and 17. The agreement for the two different calculations is seen to be quite good, especially when it is realized that the ideal diameters are proportional to the square root of the cooling times, so that if these data were plotted against bar diameters in inches, instead of time in seconds, the agreement would appear much better (nor would this be misleading, it would actually correspond in reality to quenched bar sizes).

CHANGING THE SEVERITY OF QUENCH

These calculations may now be used as an estimate of what happens when the severity of quench is changed, first by direct calculation, and then by an indirect calculation involving the equivalent diameter.

The direct calculation is illustrated in Fig. 18. It will be recalled that in the first test the quenching severity was $H = 1.6$. The

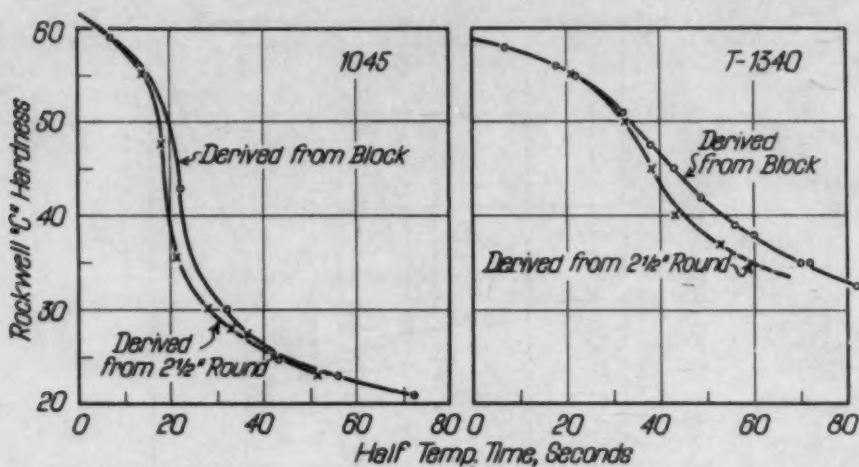


Fig. 16—Hardness-Time Relationship, Derived from Rounds and Blocks, 1045 and T-1340 Steels.

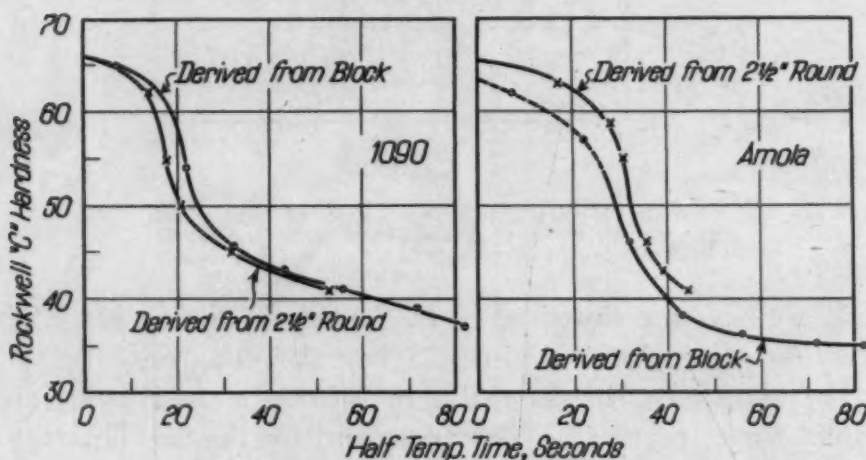


Fig. 17—Hardness-Time Relationship, Derived from Rounds and Blocks, 1090 and Amola Steels.

cooling times for this quench, and the corresponding Rockwell-C hardnesses for steel T-1340, are shown in the upper half of Fig. 18. It is seen that the lines of equal hardness at the left (isosclers) are quite the same as the lines of equal time at the right (isochrones). The new quench, $H = 1.0$, is shown in the lower half of Fig. 18, again showing both the calculated cooling times and the actual resulting hardnesses for the T-1340 steel. Naturally the cooling times at the same positions are found to be longer in the milder quench, and the resulting hardnesses are lower. But another feature of this diagram becomes apparent upon more careful examination, which may be contrasted with the previously described behavior of plates and rounds. It was pointed out that in flat plates and round bars the

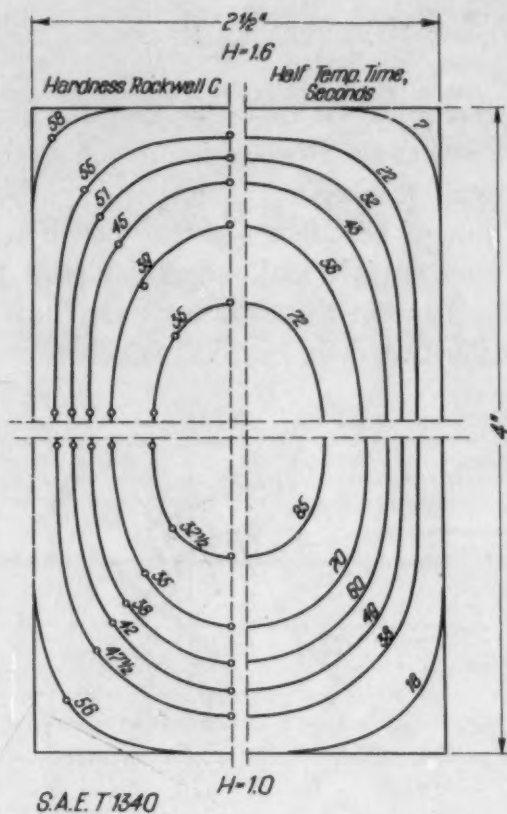


Fig. 18—Lines of Equal Cooling-Time and Lines of Equal Hardness, in Blocks Quenched $H = 1.6$ and $H = 1.0$.

heat flow (in quenching) takes place along straight lines perpendicular to the surface. Therefore in plates the lines of equal cooling time (isochrones) are straight lines parallel to the surface, and in round bars they are concentric circles. When the severity of quench is changed, the shape of these lines remains unchanged in plates and rounds; only the position changes for a particular cooling time. In blocks (and in irregular shapes) the shape of the lines changes. This may be observed in Fig. 18, for the quench $H = 1.6$ in the upper half, compared to $H = 1.0$ in the lower half. The lines of equal time were chosen so they would match on the horizontal axis (same distance from the center). But upon following these lines around to the vertical axis, it is found that in the upper half these lines are farther from the center than the corresponding lines in the lower half. Thus the curves become more rounded (approach circles) as the quench becomes milder.

EQUIVALENT DIAMETER IN RELATION TO BLOCKS

In calculating equivalent diameter for a point in a plate, see Fig. 10, the point was at the 0.6 position in the straight line from the center to the surface. Looked at in another way, the point was at the 0.6 position on a line of heat flow (in the quench), for as mentioned before these lines are straight and perpendicular to the surface. But in blocks the lines (except those at the axes) are not straight but curved. This is illustrated in Fig. 19, which represents a quarter-

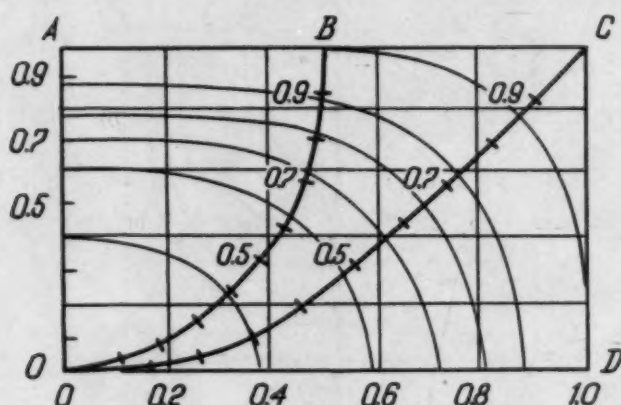


Fig. 19—Manner of Drawing "Location Lines."

section of the block, corresponding to those in Fig. 12. The six light lines are lines of equal hardness or lines of equal cooling time (in the quench). It will be realized that the lines of heat flow will always cross these lines at right angles. Four such lines of heat flow are indicated in Fig. 19 extending from the origin to points A, B, C and D. Those on the axes, the lines to A and D, are of course straight, those to B and C are curved as shown. It is of quite sufficient accuracy to sketch these lines in by hand. These heat-flow lines are the lines on which positions are located, and may be called location lines. In order to deal with specific positions, each of these "location lines" is divided into ten parts, thus suggesting the positions 0.1, 0.2, 0.3, etc. on each of the lines. These positions are dealt with in the same manner as the 0.6 position in Fig. 10, finding the equivalent diameter at each such position that is investigated. In the following calculation, the 0.5, 0.7 and 0.9 positions on each of the lines A, B, C and D are investigated. It will be realized that similar positions on the different location lines do not by any means have the same equivalent diameters: thus the 0.7 position on line A has an equivalent

diameter of 3.5 inches, whereas the 0.7 position on line C has an equivalent diameter of 2.4 inches.

The equivalent diameters may be calculated in blocks in two ways, (a) from the calculated cooling times, and (b) from the actual Rockwell hardnesses.

(a) *Equivalent Diameters from Cooling Times*—To illustrate

Table VII
Equivalent Diameter for 0.7 Point on Line B

Time t is 46 seconds.

Therefore from Fig. 15 the ideal diameter $D_i = 2.9$ inches.

Then from Fig. 3, for $D_i = 2.9$ and $H = 1.6$, read $D_o = 2.3$ inches.

Since $H = 1.6$ and $D_o = 2.3$, $HD_o = 1.6 \times 2.3 = 3.6$.

In Fig. 2, begin at the value 3.6 at the base line, and follow the curve upward to the right until it intersects the horizontal for the 0.7 position. The vertical at this point is 5.5, which represents the equivalent diameter multiplied by the quenching severity, or HD_{eq} .

Then the equivalent diameter $D_{eq} = \frac{HD_{eq}}{H} = \frac{5.5}{1.6} = 3.4$.

Table VIII
Determination of Equivalent Diameter from Calculated Half-Temperature Times
for $H = 1.6$

Heat Flow Line	Du		D_i	D_o	HD_o	HD_{eq}	D_{eq}
	D	t					
A	0.0	82	3.8	3.1	3.1
	0.5	65	3.4	2.9	4.6	5.5	3.4
	0.7	45	2.8	2.1	3.4	5.6	3.5
	0.9	19	1.8	1.2	1.9	5.6	3.5
B	0.5	65	3.4	2.9	4.6	5.5	3.4
	0.7	46	2.9	2.3	3.6	5.5	3.4
	0.9	21	1.9	1.3	2.1	6.2	3.9
C	0.5	55	3.2	2.5	4.0	4.8	3.0
	0.7	29	2.3	1.7	2.6	3.9	2.4
	0.9	8	1.2	0.7	1.1	2.5	1.6
D	0.5	65	3.4	2.9	4.6	5.5	3.4
	0.7	50	3.0	2.4	3.8	5.9	3.7
	0.9	23	2.0	1.4	2.2	7.0	4.4

the calculation of equivalent diameter from cooling time, consider the 0.7 point on location line B in Fig. 19, where the severity of quench is assumed to be $H = 1.6$. By interpolation on Fig. 18, or by calculation as in the Appendix, it is found that the half-temperature cooling time at this position is 46 seconds. From this point on, the details of the calculation are as given in Table VII. The calculations made in this manner give the equivalent diameters for the various points as shown in Table VIII.

It is now possible to make the same comparison for prisms as was made in Fig. 11 for plates, namely the relation between the re-

sults of direct calculation for a new quench and those of indirect calculation through equivalent diameters. An example of the indirect calculation is given in Table IX, and the results are tabulated in Table X, for a quenching severity $H = 1.0$.

The values given in Table X are summarized as the dashed lines of equal cooling time (isochrones) in Fig. 20, and may be compared

Table IX
Half-Temperature Time, $H = 1.0$, for 0.7 Point on Line B

From Table VIII, equivalent diameter $D_{eq} = 3.4$.
Then $HD_{eq} = 1.0 \times 3.4 = 3.4$.
On Fig. 2, where vertical 3.4 intersects horizontal 0.7, read down along curve to base line, reading value $HD_o = 2.3$.
Then $D_o = \frac{HD_o}{H} = \frac{2.3}{1.0} = 2.3$.
On Fig. 3, at $D = 2.3$ and $H = 1.0$, read $D_I = 3.2$ inches.
On Fig. 15, at $D_I = 3.2$, read $t = 57$ seconds.

Table X
Derivation of Half-Temperature Times for $H = 1.0$

Heat Flow Line	Du						
	D	D_{eq}	HD_{eq}	HD_o	D_o	D_I	t
A	0.0	3.1	4.0	89
	0.5	3.4	...	2.9	2.9	3.7	76
	0.7	3.5	3.5	2.4	2.4	3.2	57
	0.9	3.5	3.5	1.4	1.4	2.1	25
B	0.5	3.4	3.4	2.9	2.9	3.7	76
	0.7	3.4	3.4	2.3	2.3	3.2	57
	0.9	3.9	3.9	1.5	1.5	2.4	31
C	0.5	3.0	3.0	2.6	2.6	3.4	64
	0.7	2.4	2.4	1.7	1.7	2.6	37
	0.9	1.6	1.6	0.85	0.85	1.6	14
D	0.5	3.4	3.4	2.9	2.9	3.7	76
	0.7	3.7	3.7	2.5	2.5	3.3	60
	0.9	4.4	4.4	1.6	1.6	2.5	34

with the lines calculated directly, the solid lines in Fig. 20 for the same cooling times. The agreement is seen to be quite good, but it should be emphasized that the method of equivalent diameters is not intended to replace direct calculation. It is developed here to show that it can be used with reasonable accuracy in shapes too complicated for direct calculation.

(b) *Equivalent Diameters from Actual Rockwell Hardnesses*—The data of Figs. 13 and 14 may, by the use of Figs. 2, 3 and 15, be converted to the data of Table XI. Furthermore the data of Fig. 14 for the T-1340 steel may, by the use of Figs. 2 and 3, be converted to the center hardness curves of Fig. 21. These data, especially Fig. 21,

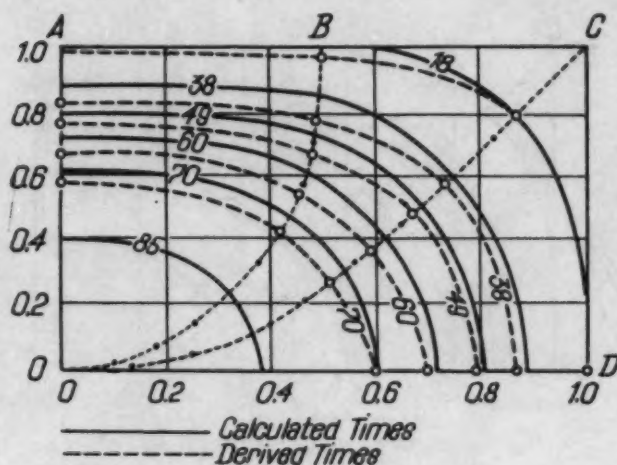
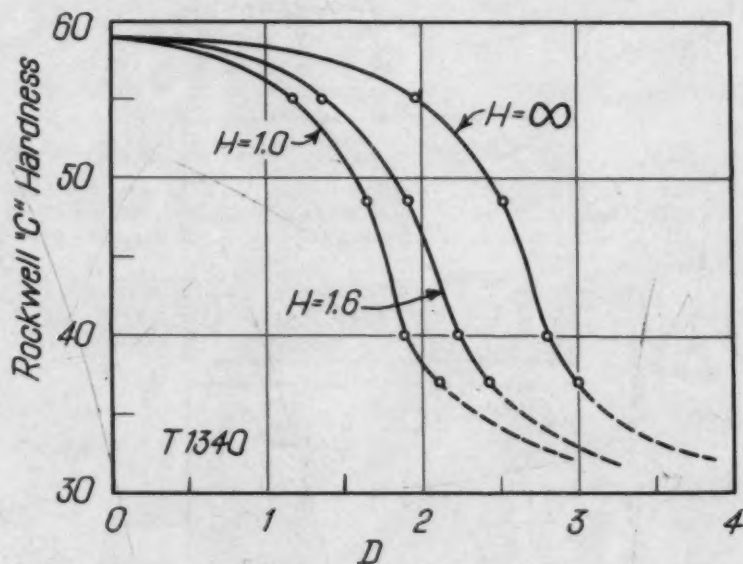


Fig. 20—As Given on Diagram.

Fig. 21—Center Hardness Curves for Three Quenches, $H = 1.0$, $H = 1.6$ and H Infinite.

make it possible to derive the equivalent diameters directly from the hardness readings.

It was found for the T-1340 steel, quenched $H = 1.6$, that the hardness at the 0.7 position on location line B was 42 Rockwell-C. From this fact the equivalent diameter may be calculated as in Table XII. The values for the various points are tabulated in Table XIII.

These equivalent diameters may now be used to estimate the Rockwell hardnesses that would be found when using some other severity of quench. For a new severity of quench $H = 1.0$, the

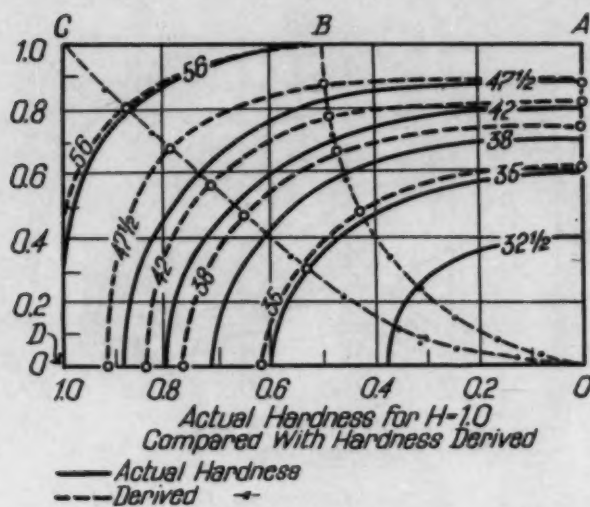


Fig. 22

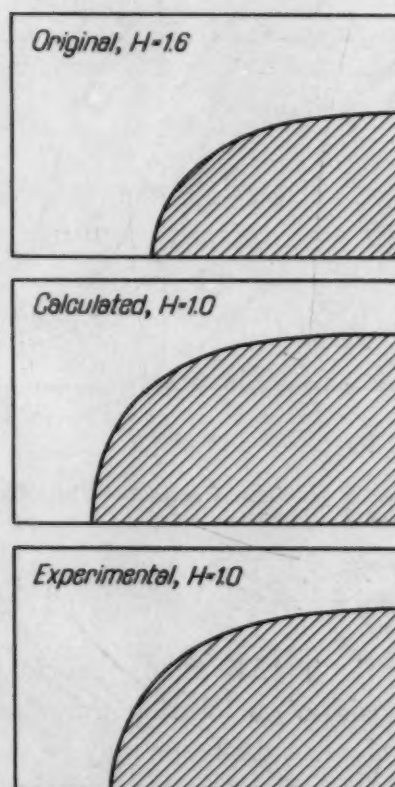


Fig. 23—Unhardened Core on Quarter Section of Quenched Blocks, Actually in Original Quench, and Calculated and Actual in New Quench.

Table XI

Steel	Maximum Theoretical Hardness 100% Martensite	Critical Hardness 50% Martensite	Critical Half-Temp. Time Derived from 2½-Inch Round	Critical Size		
				H = 1.0	H = 1.6	H = ∞
1045	62 Re	43 Re	19 seconds	1.10"	1.25"	1.85"
1090	66	54½	18	1.05	1.20	1.80
T-1340	59	39½	46	1.95	2.25	2.85
Amola	65½	51	28	1.40	1.65	2.25

Table XII

Equivalent Diameter from Hardness, at 0.7 Position on Line B

Hardness at this point is Re 42.

Then from Fig. 21, on line H = 1.6, read D_o = 2.1.Then HD_o = 1.6 × 2.1 = 3.4.Then on Fig. 2, begin at base line at 3.4 and follow curve upward and to the right until it intersects the horizontal 0.7. Read here the vertical value HD = 5.3. This is the equivalent diameter D_{eq} multiplied by the severity of quench H.

$$\text{Then } D_{eq} = \frac{HD_{eq}}{H} = \frac{5.3}{1.6} = 3.3.$$

Table XIII

Determination of Equivalent Diameters from Actual Hardness Values for H = 1.6

Heat Flow Line	Du					
	D	Re	D _o	HD _o	HD _{eq}	D _{eq}
A	0.0	33	3.0	3.0
	0.5	37	2.5	3.9	4.6	2.9
	0.7	44	2.1	3.4	5.1	3.2
	0.9	56	1.2	2.0	5.7	3.6
B	0.5	36½	2.5	4.0	4.8	3.0
	0.7	42	2.1	3.4	5.3	3.3
	0.9	54	1.5	2.4	7.5	4.7
	0.5	40	2.2	3.6	4.3	2.7
C	0.7	51½	1.7	2.7	4.0	2.5
	0.9	58	0.75	1.2	2.8	1.7
	0.5	36½	2.5	4.0	4.8	3.0
	0.7	42	2.1	3.4	5.3	3.3
D	0.9	54	1.5	2.4	7.5	4.7

calculation for the 0.7 point on line B is carried out in Table XIV, and the calculations for the various points are summarized in Table XV.

It is now finally possible to collect these values as in the dashed lines in Fig. 22, and compare them with the values actually found by experiment, shown as the solid lines in Fig. 22. The agreement is again seen to be quite fair.

An idea of the effect of change of quench, and also the order of accuracy, may be gained from Fig. 23, which sketches the unhard-

Table XIV
New Rockwell Hardness Estimated for $H = 1.0$ at 0.7 Position on Line B

Equivalent diameter $D_{eq} = 3.3$ (from Table XIII).

Then $H_{Deq} = 1.0 \times 3.3 = 3.3$.

On Fig. 2, where vertical line 3.3 crosses horizontal line 0.7, from this intersection follow curve downward to the left to base line and read $HD_0 = 2.2$

Then $D_0 = \frac{HD_0}{H} = \frac{2.2}{1.0} = 2.2$.

On Fig. 21, on curve for $H = 1.0$, at $D = 2.2$, read Rockwell C = 36.

This is the hardness at this point in the new quench.

Table XV
Derivation of Hardness Values for $H = 1.0$ from Equivalent Diameters

Heat Flow Line	Du					
	D	D_{eq}	H_{Deq}	HD_0	D_0	R_c
A	0.0	3.0	32
	0.5	2.9	2.9	2.5	2.5	35
	0.7	3.2	3.2	2.2	2.2	37
	0.9	3.6	3.6	1.5	1.5	52
B	0.5	3.0	3.0	2.6	2.6	34
	0.7	3.3	3.3	2.2	2.2	36
	0.9	4.7	4.7	1.7	1.7	47
	0.5	2.7	2.7	2.3	2.3	36
C	0.7	2.5	2.5	1.8	1.8	43
	0.9	1.7	1.7	0.9	0.9	57
	0.5	3.0	3.0	2.6	2.6	34
	0.7	3.3	3.3	2.2	2.2	36
D	0.9	4.7	4.7	1.7	1.7	47

ened cores as originally quenched at $H = 1.6$, and as calculated and actually obtained for the new quench $H = 1.0$. The increase of unhardened core in the milder quench was thus calculated to be somewhat greater than obtained in reality, but the diagrams show that a reasonable estimate was obtained.

Precisely similar principles may be employed for irregular shapes. Thus Fig. 24 shows a section through a journal and pin portion of a crankshaft that was investigated. Treated in a way similar to Fig. 19, the light lines in Fig. 24 are lines of equal hardness, and the heavy lines are location lines. The latter are sketched in by hand, so that they always cross the equal-hardness lines at right angles, and each location line is then divided into ten parts as shown. The previously outlined procedure may then be followed precisely.

SUMMARY

For the estimation of hardenability and quenching effects in flat plates, a method of direct calculation is provided, analogous to the system suggested previously for round bars.

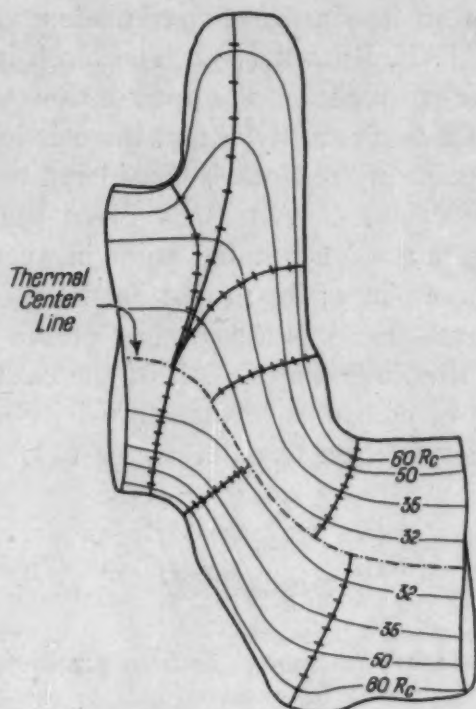


Fig. 24—Manner of Placing "Location Lines" on Irregular Shape.

For the study of shapes too irregular for calculation, an approximation by the method of "equivalent diameters" is suggested.

ACKNOWLEDGMENTS

Great appreciation is expressed for the co-operation of members of the metallurgical staffs of the Carnegie-Illinois Steel Corporation, and especially W. F. Craig, who made many of the calculations.

References

1. M. A. Grossmann, M. Asimow and S. F. Urban, in volume "Hardenability of Alloy Steels," American Society for Metals, p. 124.
2. M. A. Grossmann and M. Asimow, *Iron Age*, 1940, April 25, p. 25 and May 2, p. 39.
3. T. F. Russell, First Report of Alloy Steels Research Committee, Iron and Steel Institute (British), 1936, p. 149-187.
4. G. V. Luerssen, in volume "Hardenability of Alloy Steels," American Society for Metals, p. 302.
5. J. B. Austin, U. S. Steel Corporation Research Laboratory, private communication.

APPENDIX

The calculations of half-temperature times for a rectangular

block, cooled on all of its six faces, were made according to the suggested method of T. F. Russell (3). The block is considered to be the enclosed space formed by the intersection of three pairs of parallel planes which individually bound three infinite plates.

A particular point in the block is considered to belong separately to each of the three plates. After any given passage of time, the proportional temperature level at the point in question is calculated for each of the three plates; the proportional level for the block for this same point is calculated as the product of the proportional temperature levels in the three plates. All of the calculations are based on the assumption of constant thermal diffusivity and Newton's law of cooling with constant heat transfer factor (1).

DISCUSSION

Written Discussion: By Arthur E. Focke, research metallurgist, Diamond Chain and Manufacturing Co., Indianapolis, Ind.

In the past, while hardenability and the various hardenability tests have been of great academic interest to me, it has never seemed desirable to attempt to apply such tests to the solution of our practical problems.

In the first place, the difference between the quench which the part would receive in production and that which would be applied in the laboratory seemed to offer a serious limitation.

Secondly, it did not seem possible to apply the information obtained from hardening bars, except in a general way, to irregular objects such as sprockets, with which we are concerned.

Finally, it has been possible at least for the last ten years, whenever it was necessary, to specify and obtain readily, steels with sufficient hardenability to insure satisfactory parts.

The paper on "Relation of Hardenability to Quenching" by Mr. Asimow and Dr. Grossmann presented before this Society in 1938 showed that my first objection could be overcome and this paper indicates that the second is not insurmountable. Also, we have to recognize that, at least for the present, it is no longer possible always to obtain the desired analysis in a reasonable time and it is necessary to consider substitutes. The necessity for meeting rigid specifications in the finished part and providing prompt delivery makes it of paramount importance that it is known that the finished result will be acceptable before such substitutes can be authorized.

The logical answer appears to be a hardenability test and if it is possible, as suggested in this paper, to predict from tests made on simple billet disks the results which may be expected in our finished part, our problem of specifying materials and treatments will be greatly simplified.

It is for these reasons that I feel that this paper is particularly valuable.

Written Discussion: By C. B. Post and O. V. Greene, The Carpenter Steel Company, Reading, Pa.

The effect of shape as a factor in hardenability has no direct effect upon quenching, but has a very important indirect effect on the final results. This indirect effect is due to the variation in the rate of heat extraction because of differences in the contour of surfaces. This characteristic has engaged the attention of a few investigators, notably French¹ and Luerssen², and naturally any addition to the literature on this subject, as exemplified by the paper of Messrs. Asimow and Grossmann, is valuable.

In attempting to apply the authors' method of analysis to five heats of 1.10 per cent carbon shallow hardening steel, we found that predicted values of case depth in plates or disks, based on information obtained on $\frac{3}{4}$ -inch round specimens, showed no correlation with observed data on the disks. We have shown in discussion of the paper by Messrs. Grossmann and Stephenson in this session, entitled "The Effect of Grain Size on Hardenability", that the quenching severity of rounds of 1.10 per cent carbon shallow hardening steel, quenched in a vertical 3-inch diameter flush, is more nearly represented by $H = 5$ than $H = 10$ or $H = 12$ (See Fig. 1 of discussion mentioned above). Furthermore, the use of the standard flushing fixture in the Shepherd hardenability test³ actually results in a much more drastic quench than $H = 12$. Consequently we do not believe it is reasonable to use the same quenching severity on both disks and rounds as was used by the authors in analyzing information presented by Mr. Luerssen². Evidently the agreement found by Messrs. Asimow and Grossmann between their predicted penetrations and the aforementioned data of Mr. Luerssen is fortuitous.

Considering the severity of quench characteristic of the standard flushing fixture in the Shepherd hardenability test³, the data obtained from the above five different heats of 1.10 per cent carbon shallow hardening steels shows this severity to be in fact greater than $H = 12$, and for the purpose of calculations, a severity of quench represented by $H = 15$ has been found to be applicable to this type of flush. In reading the charts presented by Messrs. Asimow and Grossmann this difference in severity of quench between $H = 12$ and $H = 15$ is negligible.

Rather than accepting such a theoretical relationship as found by Messrs. Asimow and Grossmann, i.e., $D_1 = 1.377 L_1$, we have preferred to evaluate data from ten heats of 1.10 per cent carbon shallow hardening tool steels in order to determine a conversion factor based upon experimental data. The data obtained from these heats include penetrations in 64ths of an inch on $\frac{3}{4}$ -inch round specimens which were oil-treated 1600 degrees Fahr.—40 minutes at heat, and brine-quenched 1450 degrees Fahr., and penetrations on disks ranging in thickness from $\frac{3}{8}$ to $\frac{1}{2}$ inch. These disks were oil-treated 1600 degrees Fahr.—40 minutes at heat, and brine-quenched 1425 to 1435 degrees Fahr. The critical bar diameter was computed from the penetration obtained

¹H. J. French, "A Study of the Quenching of Steels," *TRANSACTIONS, American Society for Steel Treating*, Vol. XVII, 1930, p. 646.

²G. V. Luerssen, "Hardenability in Light Sections," *Hardenability of Alloy Steels*, published by American Society for Metals, 1939, p. 302.

³B. F. Shepherd, "Inherent Hardenability Characteristics of Tool Steel," *TRANSACTIONS, American Society for Steel Treating*, Vol. XVII, 1930, p. 90.

on the $\frac{3}{4}$ -inch round specimens (using the recommended procedure of Dr. Grossmann, et al., and a severity of quench of $H = 5$). From this critical bar diameter, the so-called "ideal" critical bar diameter was obtained. From the penetrations obtained on the disks, the critical disk thickness was computed (using the procedure recommended in this paper and a severity of quench of $H = 15$). From this latter value the so-called "ideal" critical disk thickness was obtained.

The results obtained from these ten heats are shown in Fig. 1. This figure shows the experimental correlation between "ideal" critical bar diameter (obtained from the $\frac{3}{4}$ -inch round specimen and $H = 5$), and "ideal" critical

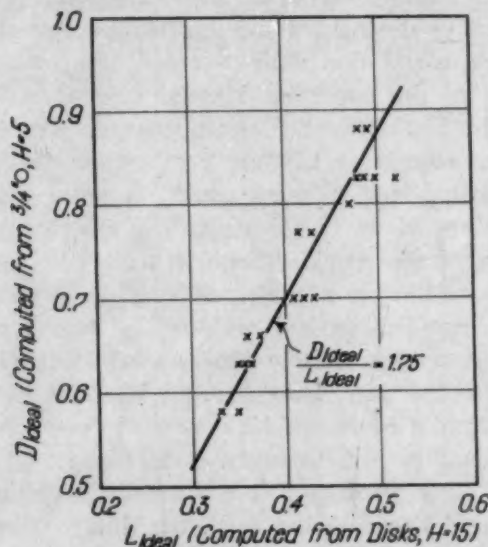


Fig. 1—Experimental Correlation Between "Ideal" Critical Diameter, Obtained From $\frac{3}{4}$ Inch Diameter and $H = 5$, and "Ideal" Critical Slab Thickness, Obtained From Disks and $H = 15$.

slab thickness (obtained from hardenability disks and $H = 15$). These data, shown in Fig. 1, indicate clearly that the theoretical relationship suggested by Asimow and Grossmann, i. e., $D_1 = 1.377 L_1$, should be corrected to read $D_1 = 1.75 L_1$.

Incidentally, French⁴ determined experimentally the center cooling velocity of various simple shapes as determined by the exposed surface per unit of volume. We quote from page 45 of Mr. French's book:⁴ "When quenching from a definite temperature into a given coolant, the center cooling velocity is determined by the exposed surface per unit of volume. Hence, it will be equal in spheres, rounds, and plates, which have equal surface per unit of volume. This condition is fulfilled when the ratio of diameters of spheres and rounds of infinite length to the thickness of plates having infinite width and length is 3:2:1. When the length of the round is four times its diameter and the width and length are each four times the thickness of the plate, these

⁴H. J. French, "The Quenching of Steels," published by American Society for Steel Treating, 1930, p. 45.

Table I

Heat No.	Penetration in $\frac{3}{4}$ -Inch Round	D_u D	Disk Thickness Inches	Core Thickness	Derittical (H = 5)	D_{ideal}	L_{ideal}	Derittical (H = 15)	Predicted Core Thickness
A	10/64	0.584	12/32 13/32	0 0.109-0.125	0.59	0.77	0.44	0.37	0 0.140
B	12/64	0.499	12/32 13/32	0 0.125	0.62	0.79	0.45	0.38	0 0.140
C	10.5/64	0.563	11/32 12/32 13/32	0 0.125 0.171	0.60	0.77	0.44	0.37	0 Trace 0.140
D	8/64	0.666	10/32 11/32 12/32 13/32	0 0.125 0.171 0.203	0.53	0.69	0.394	0.33-0.34	0 0.085 0.168 0.215
E	9/64	0.625	11/32 12/32 13/32	0 0.109 0.156	0.56	0.71	0.405	0.35	0 0.109 0.185
F	6.5/64	0.73	.300 10/32 11/32	0 0.093 0.156	0.48	0.65	0.372	0.31	0 0.060 0.145
G	6/64	0.75	9/32 .300 10/32 11/32	0 0.078 0.093 0.140	0.46	0.62	0.354	0.29	0 0.060 0.103 0.170
H	17/64	0.293	12/32 13/32 14/32 15/32 16/32	0 0 0.140-0.171 0.187-0.218 0.250	0.70	0.88	0.503	0.43	0 0 0.100 0.170 0.190

ratios become 4:3:2." In other words, the value determined by us from experimental data, i.e., $D_1 = 1.75 L_1$, is midway between the infinite condition represented by a ratio of 2:1 for the round to the plate, and 3:2 for a condition wherein the length of the round is four times its diameter, and the width and length of the disk are four times its thickness. This is only one more instance of the inherent exactness to be found in the painstaking, fundamental researches of H. J. French.

P-F hardenability data from eight heats of 1.10 per cent carbon shallow hardening tool steels were analyzed by means of the recommended procedure of Messrs. Asimow and Grossmann, but with the correction of $D_1 = 1.75 L_1$. The experimental data included penetrations in 64ths of an inch obtained on $\frac{3}{4}$ -inch round specimens when oil-treated from 1600 degrees Fahr.—40 minutes at heat, and brine-quenched 1450 degrees Fahr., and the penetration on disks ranging from $\frac{1}{2}$ to 16/32 inch in thickness when oil-treated from 1600 degrees Fahr.—40 minutes at heat, brine-quenched 1425 to 1435 degrees Fahr. in the standard Shepherd hardenability fixture³. These data are shown in columns 2, 3 and 4 of Table I. From the penetration obtained on the $\frac{3}{4}$ -inch round specimens, the critical bar diameter may be computed by means of the recommended procedure of Grossmann using a severity of quench of $H = 5$. From this latter value the "ideal" critical bar diameter may be obtained, and these are shown in column 7, Table I. The relationship $D_1 = 1.75 L_1$ may be used to compute the "ideal" critical disk thickness, and these values are given in column 8, Table I. Use may then be made of authors' Fig. 6 to obtain the

critical disk thickness for a severity of quench of $H = 15$. Likewise, authors' Fig. 4 may be used to estimate the core thickness for these theoretical values.

Referring now to Table I, it is of interest to note that the predicted disk thickness which will just harden through agrees in every case with the observed data. This latter conclusion follows from an inspection of column 10 and 5, Table I. There is fair agreement between the predicted core thicknesses for the disks having a greater thickness than the critical section, as may be ascertained by consulting again columns 10 and 5, Table I. It might be well to point out at this time that if $H = 12$ were used for the severity of quench appropriate to the $\frac{3}{4}$ -inch round data, large disagreement would be encountered between the predicted and experimental critical disk thickness.

The extrapolation of data obtained from simple sections, such as rounds, spheres, and flats, to complex shapes is a question which depends entirely upon the complexity of shape under investigation. In addition to this, there is also the possibility that for a given coolant, the severity of quench will vary for different size sections, and different shapes. This is to say, if the severity of quench for a given coolant has been determined by measurements made in the Grossmann manner on a series of different size round bars, what assurance have we that this same severity of quench, expressed by a definite H value, will hold when plates, blocks, gears (especially with deep recesses), and other irregular shapes, are quenched in this same coolant? The answer to this question does not involve a consideration of Grossmann's method of analysis, and it might well be that in order to determine the severity of quench appropriate to a given section, this severity should be determined on some section approximately similar in its behavior. It should be understood that we do not mean to involve section as one of the factors of quenching severity, but are merely pointing out that the effect of mass and surface may affect the rate of heat extraction and consequently show an apparent effect upon the quenching severity. This is particularly true if very large, irregular shapes are quenched in media of insufficient volume.

Written Discussion: By W. E. Jominy, research laboratories division, General Motors Corp., Detroit.

The authors have presented an ingenious method for calculating the expected hardness at any point in a hardened section. The method depends fundamentally on the assumption that with a given steel section all points having the same cooling rate or, as the authors prefer to state, the same "half temperature time", will be of the same hardness. This assumption appears quite sound provided of course that the steel has a uniform composition. If we do not have uniform composition inaccuracies will no doubt result.

Probably what is apt to be of a more serious nature is the difficulty of maintaining a uniform quench in making the immersion type of hardenability test required in this method of measurement. The method assumes that a given severity of quench is repeated exactly with different sized bars and sections by the investigator who is to make the hardenability calculations. It has been my observation that this is seldom attained in any type of "hand" quenching—especially with the less severe quenching media like oil.

In making cooling rate determinations by hand quenching in oil it has

been our experience that somewhat different rates result with each attempt and that a mechanical means of circulating the oil around the piece is necessary to obtain uniform results. Furthermore, with certain steels, the effect of scale is quite large so that we would not expect to get uniform results unless we prevent the formation of scale.

This lack of uniformity in quenching, I believe, leads to difficulty in attempting to determine the severity of quench.

I have tried to determine the ideal critical size by the method of Asimow and Grossmann on a number of steels whose cross sectional hardness have been published in the technical literature, and in most cases have had difficulty in ascertaining the severity of quench. The U curves I studied in this manner have been those published by Burns, Moore and Archer, A. J. Scheid, Jr., and Klain and Lorig. It was my experience that the Du/D points plotted against the diameters in a series of various sized rounds for a given steel did not fall on any of the "identical cooling time" curves shown in Fig. 2. As a result it was not possible to determine the H value for the severity of quench, which is necessary to determine the critical size.

For example, if we consider the U curves for water quenched S.A.E. 1035 steel published by Burns, Moore and Archer, we find that the Du/D values for the five sizes, plotted against their respective diameters do not coincide with any curve on Fig. 2. If we use only the $\frac{3}{4}$, $1\frac{1}{2}$ and 2-inch rounds we find $H = 16$. If we use only the points for the 1, $1\frac{1}{2}$ and 2-inch rounds, the H value appears to be 3.4. If we take the value $H = 16$ then the critical size for this steel would be 0.62 inches and the ideal critical size 0.68 inches. If we use the value of $H = 3.4$ then the critical size would be 0.77 inches and the ideal critical size 1.05 inches.

While possible differences in hand quenching might explain the errors in attempting to calculate the severity of quench, H , from Burns, Moore and Archer's curves, it should be remembered that their method gave satisfactory duplication of results in hardenability measurement. The curves of Klain and Lorig were produced by mechanical control of the flow of the quenching medium past the test specimen and these likewise do not yield to the method of Asimow and Grossmann for determining severity of quench. There are no doubt other factors, some of which were discussed by the authors in their previous paper of 1938, that affect these results.

When it comes to oil-quenched bars, the determination of H is much more difficult and in my experience is seldom feasible by the method suggested.

I have never found a set of U curves with oil-quenched steel whose points of equal hardness gave Du/D values that coincided with the curves of Fig. 2.

These observations would seem to indicate that much care should be exercised in quenching bars for the determination of the severity of quench value H and that it would be better to determine the critical bar size experimentally than by calculation.

It is the writer's experience that the method is not so well suited to oil hardening steels and best suited to steels requiring severe quenching operations.

Authors' Reply

We appreciate these discussions very much. We should like to thank Dr. Focke for his remarks and cordial comments on the methods. We certainly agree with him that it is very necessary in any of these considerations to focus attention separately on the severity of quench and the hardenability of the steel, and to bear in mind separately the values so derived.

With regard to the discussion by Messrs. Post and Greene, they bring up a number of points. In conformity with their very thorough analysis, we want to say first that we consider they are very right about this uncertainty of having the same severity of quench on different shapes and different size pieces quenched ostensibly alike. In fact, it should be borne in mind when quenching a complex shape, even if we thought we knew what we might call the severity of quench for that shape, it must be realized that the severity varies at different positions on the shape. For example, the severity of quench is likely to be greater on a projecting part and less at a re-entrant angle.

We confess to some surprise at finding their relationship 1.75 for the ideal diameter versus the ideal plate thickness. Certainly there can be no question of their data because all of their figures fall very beautifully on the line 1.75. We feel that a great deal of thought will have to be given to this point and would like to study their data in more detail to find out where the discrepancy occurs. We are inclined to believe that the other figure given in this paper has justification in that it represents an ideal quench. It remains to be seen what the actual quenching conditions are which affect this number.

In regard to French's work, we certainly concur in the praise which the authors have given to this very careful work. A point to be borne in mind is that the relationship between diameter and plate varies with the severity of quench. That is to say, if in the ideal quench the relationship factor is either 1.75 or 1.37, that would not be the relationship for some other quench.

Further, in regard to the differences in severity of quench with different shapes, there was some mention of the departure from Newton's law of cooling, which was the basis of necessity of Mr. Asimow's calculations on heat flow. In addition to the vapor film which forms, which causes some departure from Newton's law of cooling, there is also the matter of turbulence of the liquid as determined by the shape of the piece, the manner in which it is agitated, and the manner in which the cooling liquid flows in the tank.

Well, all of this perhaps may serve only to frighten experimenters away from any attempt at estimating severity of quench. Nevertheless, we must say with some positiveness that we have found estimations of severity of quench to be very useful in giving general broad trends. It is quite true that small differences arise and it is particularly for this reason that we believe that special appreciation should be expressed for the work of Messrs. Post and Greene, because they have taken the time and the trouble to provide some very accurate data.

We appreciate Mr. Jominy's calling attention to possible difficulties in measuring severity of quench. We believe that the difficulty which arises here and which we have encountered to our own regret should be explained

first by calling attention to variations which have a disastrous effect in attempting to estimate severity of quench. Consider the quenching of two bars, quenched ostensibly alike. If they are indeed quenched precisely alike, then the severity of quench judged from them will be correct. But if they are not quenched quite alike, then the estimated severity will not be an average but will unfortunately be misleadingly and exaggeratedly too large or too small. Thus if a 1-inch and a 1½-inch bar are supposed to be quenched alike, but the 1-inch bar is quenched with a severity $H = 2$, and the 1½-inch bar is quenched with a severity $H = 2.5$, the apparent severity will not be the average $H = 2.25$, but the exaggerated value $H = 4.1$. An examination of the curves will reveal why this is so. Well now, we cannot by any means expect that the quenches will be absolutely identical on different bars. Therefore, we regretfully come to the conclusion that we have to use a considerable number of bars to get a good average value, and we now recommend using at least four and preferably more sizes when attempting to estimate severity of quench.

With regard to discrepancies, we should like to say that we believe very detailed attention must be paid to the cases to be analyzed. For instance, Mr. Jominy cited the case reported by Klain and Lorig. It will be noted in that case, which we too were unsuccessful in analyzing completely, that the larger sizes were quenched from a higher temperature than the smaller sizes, which for all we know may have introduced changes in grain size, changes in amount of scaling, or what not. It is extremely important to have identical steels, identically heat treated, and in view of the difficulties involved, treat a number of sizes.

**TECHNICAL PROGRAM AND REPORTS OF OFFICERS
AMERICAN SOCIETY FOR METALS—22nd ANNUAL
CONVENTION, CLEVELAND, OCTOBER 21 to 25, 1940**

FOR the benefit of members who were not in attendance at the Twenty-second Annual Convention of the Society, held in Cleveland, October 21 to 25, 1940, and for purposes of record, the Technical Papers Program and Reports of Officers presented at the annual meeting are herewith published in full.

MONDAY, OCTOBER 21

Morning Session—Grand Ballroom, Hotel Statler

Joint Chairmen—H. E. Brown and J. J. Kanter

Correlation of High Temperature Creep and Rupture Test Results, by R. H. Thielemann, General Electric Co.

The Development of Alloys for Use at Temperatures Above 1000 Degrees Fahr., by E. R. Parker, General Electric Co.

The Significance of Hydrogen in the Metallurgy of Malleable Cast Iron, by H. A. Schwartz, G. M. Guiler and M. K. Barnett, National Malleable & Steel Castings Co.

Factors Affecting the Activity of Carburizing Compounds, by M. Sutton, Standard Oil Co. of Indiana, and R. A. Ragatz, University of Wisconsin.

Afternoon Session—Public Auditorium—Ball Room—4th Floor

Joint Chairmen—D. F. McFarland and H. W. McQuaid

A Balanced Protective Atmosphere—Its Production and Control, by J. R. Gier, Westinghouse Electric & Mfg. Co.

Furnace Atmosphere Generation, by Sam Tour, Lucius Pitkin, Inc.

Dimensional Changes on Hardening High Chromium Tool Steels, by H. Scott and T. H. Gray, Westinghouse Electric & Mfg. Co.

Water Vapor in Furnace Atmospheres, by Sam Tour, Lucius Pitkin, Inc.

Public Auditorium—Club Room B—3rd Floor

Educational Lectures

Educational Course on *The Strength of Metals Under Combined Stresses*, by Maxwell Gensamer, Carnegie Institute of Technology.

Public Auditorium—Ball Room—4th Floor

Educational Course on *Quenching of Metals: Principles of Quenching*, by A. Allan Bates, Westinghouse Electric & Mfg. Co.

TUESDAY, OCTOBER 22

Morning Session—Grand Ballroom, Hotel Statler

Joint Chairmen—E. C. Bain and G. T. Williams

Transformation of Austenite on Continuous Cooling and Its Relation to Transformation at Constant Temperature, by R. A. Grange, United States Steel Corp., and J. M. Kiefer, American Steel and Wire Co.

Influence of Austenitic Grain Size on the Critical Cooling Rate of High Purity Iron-Carbon Alloys, by T. G. Digges, National Bureau of Standards.

Dilatometric Studies in the Transformation of Austenite in a Molybdenum Cast Iron, by D. B. Oakley, Bethlehem Steel Co., and J. F. Oesterle, University of Wisconsin.

Effect of Rate of Heating Through the Transformation Range on Austenitic Grain Size, by S. J. Rosenberg and T. G. Digges, National Bureau of Standards.

Morning Simultaneous Session—Euclid Ballroom, Hotel Statler

Joint Chairmen—R. L. Dowdell and H. E. Handy

Recovery of Nickel From Cold Working on Annealing, by Erich Fetz, Wilbur B. Driver Co.

Structural Changes in Low Carbon Steels Produced by Hot and Cold Rolling, by N. P. Goss, Cold Metal Process Co.

Quantitative Measurement of Strain Hardness in Austenitic Manganese Steel, by D. Niconoff, Republic Steel Corp.

Equilibrium Relations in the Solid State of the Iron-Cobalt System, by W. C. Ellis and E. S. Greiner, Bell Telephone Laboratories.

Afternoon Session—Public Auditorium—Ball Room—4th Floor

Joint Chairmen—J. F. Harper and T. H. Nelson

The Effect of Molybdenum and Columbium on the Structure, Physical Properties and Corrosion Resistance of Austenitic Stainless Steels, by R. Franks, W. O. Binder and C. R. Bishop, Union Carbide & Carbon Research Laboratories.

Kinetics and Reaction Products of the Isothermal Transformation of a 6 Per Cent Tungsten, 6 Per Cent Molybdenum High Speed Steel, by J. L. Ham, R. M. Parke and A. J. Herzig, Climax Molybdenum Co.

Some Surface Studies on Treated High Speed Steel, by J. G. Morrison, Landis Machine Co.

Surface Carbon Chemistry and Grain Size of 18-4-1 High Speed Steel, by W. A. Schlegel, The Carpenter Steel Co.

Public Auditorium—Club Room B—3rd Floor

Educational Lectures

Educational Course on *The Strength of Metals Under Combined Stresses*, by Maxwell Gensamer, Carnegie Institute of Technology.

Public Auditorium—Ball Room—4th Floor

Educational Course on *Quenching of Metals: Quenching in Production Heat Treating*, by W. J. Conley, University of Rochester.

WEDNESDAY, OCTOBER 23

Grand Ballroom, Hotel Statler

Annual Meeting of the American Society for Metals

1940 Edward de Mille Campbell Memorial Lecture, by S. L. Hoyt, Battelle Memorial Institute

Chairman—Clyde E. Williams

Afternoon Session—Public Auditorium—Ball Room—4th Floor

Joint Chairmen—R. F. Mehl and W. M. Saunders, Jr.

The Technique of Microradiography and Its Application to Metals, by G. L. Clark, University of Illinois, and W. M. Shafer, Iowa State Teachers College.

Notes on the Interpretation of X-Ray Diffraction Diagrams and Evidence of Mosaic Structures, by N. P. Goss, Cold Metal Process Co.

Quantitative Evaluation of Distortion in Silicon Steel and in Aluminum, by G. L. Clark, University of Illinois, and W. M. Shafer, Iowa State Teachers College.

Effects Upon Furnace Refractories of Protective Gases for High Carbon Steels, by J. H. Loux, Salem Engineering Co.

Afternoon Simultaneous Session

Public Auditorium—Club Room B—3rd Floor

Joint Chairmen—W. E. Mahin and A. W. Winston

Further Notes of Precipitation Hardening in the Heavy Alloys, by W. P. Sykes, General Electric Co.

Some Properties of Hot-Pressed and Sintered Copper Powder Compacts, by C. G. Goetzl, American Electro Metal Corp.

The Precipitation Reaction in Cold-Rolled Phosphor Bronze; Its Effects on Hardness, Conductivity, and Tensile Properties, by R. H. Harrington and R. G. Thompson, General Electric Co.

The Effect of Impurities on Some High Temperature Properties of Copper, by E. R. Parker, General Electric Co.

Public Auditorium—Club Room B—3rd Floor

Educational Lectures

Educational Course on *The Strength of Metals Under Combined Stresses*, by Maxwell Gensamer, Carnegie Institute of Technology.

Public Auditorium—Ball Room—4th Floor

Educational Course on *Quenching of Metals: Quenching in Custom Heat Treating*, by R. G. Roshong, Lindberg Steel Treating Co.

THURSDAY, OCTOBER 24**Morning Session—Grand Ballroom, Hotel Statler**

Joint Chairmen—G. V. Luerssen and B. A. Rogers

Cementite Stability and Its Relation to Grain Size, Abnormality and Hardenability, by C. R. Austin and M. C. Fetzner, Pennsylvania State College.

Effect of Deoxidation on Hardenability, by G. V. Cash, General Electric Co., T. W. Merrill, U. S. Steel Corp., and R. L. Stephenson, Carnegie-Illinois Steel Corp.

The Effect of Grain Size on Hardenability, by M. A. Grossmann and R. L. Stephenson, Carnegie-Illinois Steel Corp.

Hardenability Characteristics of Various Shapes, by M. Asimow and M. A. Grossmann, Carnegie-Illinois Steel Corp.

Morning Simultaneous Session—Euclid Ballroom, Hotel Statler

Joint Chairmen—E. S. Davenport and M. F. Judkins

Effect of Nitrogen on the Case Hardness of Two Alloy Steels, by S. W. Poole, Republic Steel Corp.

Influence of Silicon and Aluminum Additions on the Constitutional Diagram of 4-6 Cr-Mo Steels, by C. L. Clark, Timken Roller Bearing Co., and M. A. Bredig, Vanadium Corp. of America.

The Effect of Molybdenum on the Isothermal, Subcritical Transformation of Austenite in Low and Medium Carbon Steels, by J. R. Blanchard, R. M. Parke and A. J. Herzig, Climax Molybdenum Co.

Effects of Small Amounts of Alloying Elements on the Tempering of Pure Hypereutectoid Steels, by C. R. Austin and B. S. Norris, Pennsylvania State College.

Afternoon Session—Public Auditorium—Ball Room—4th Floor

Symposium on Surface Treatment of Metals

Joint Chairmen—A. A. Bates and L. W. Kempf

Anodic Treatment of Aluminum, by J. D. Edwards, Aluminum Co. of America.

The Passivation and Coloring of Stainless Steel, by G. C. Kiefer, Allegheny Ludlum Steel Corp.

The Surface Treatment of Magnesium Alloys, by H. W. Schmidt, W. H. Gross and H. K. DeLong, Dow Chemical Co.

Corrosion Resistance of Tin Plate; Influence of Steel Base Composition on Service Life of Tin Plate Containers, by R. R. Hartwell, American Can Co.

Zinc Coatings: Unit Operations, Costs and Properties, by J. L. Bray, Purdue University, and F. R. Morral, Continental Steel Corp.

Afternoon Simultaneous Session**Public Auditorium—Club Room B—3rd Floor**

Joint Chairmen—W. R. Breeler and H. D. Churchill

The Fatigue and Bending Properties of Cold Drawn Steel Wire, by H. J. Godfrey, Lehigh University, Fritz Engineering Laboratory.

The Chafing Fatigue Strength of Some Metals and Alloys, by G. Sachs and P. Stefan, Case School of Applied Science.

Fatigue and Damping Studies of Aircraft Sheet Materials; Duralumin Alloy 24ST, Alclad 24ST and Several 18-8 Type Stainless Steels, by R. M. Brick and Arthur Phillips, Yale University.

Alloys of Manganese and Copper: Vibration-Damping Capacity, by R. S. Dean, C. T. Anderson and E. V. Potter, Bureau of Mines, U. S. Dept. of Interior.

Public Auditorium—Club Room B—3rd Floor

Educational Lecture

Educational Course on *The Strength of Metals Under Combined Stresses*, by Maxwell Gensamer, Carnegie Institute of Technology.

Grand Ballroom, Hotel Statler

Annual Banquet of the American Society for Metals

FRIDAY, OCTOBER 25**Lattice Room, Hotel Statler***Educational Lecture*

Educational Course on *The Strength of Metals Under Combined Stresses*, by Maxwell Gensamer, Carnegie Institute of Technology.

Morning Session—Grand Ballroom, Hotel Statler*Symposium on Surface Treatment of Metals*

Joint Chairmen—J. J. Crowe and J. F. Wyzalek

Diffusion Coatings on Metals, by F. N. Rhines, Carnegie Institute of Technology.

Surface Reactions and Diffusion, by J. E. Dorn, J. T. Gier, L. M. K. Boelter and N. F. Ward, University of California.

Heat Treating with Induction Heat, by Edmund Blasko, Ford Motor Co.

Inherent Characteristics of Induction Hardening, by M. A. Tran, Park Drop Forge Co., and H. B. Osborn, Jr., Ohio Crankshaft Co.

Flame Pretreatment of Structural Steel Surfaces for Painting, by J. G. Magrath, Air Reduction Sales Co.

Afternoon Session—Public Auditorium—Club Room B—3rd Floor*Symposium on Surface Treatment of Metals*

Joint Chairmen—W. P. Eddy, Jr., and J. F. Oesterle

Shot Blasting and Its Effect on Fatigue Life, by F. P. Zimmerli, Barnes-Gibson-Raymond, Division of Associated Spring Corp.

Effect of Surface Conditions on Fatigue Properties, by O. J. Horger and H. R. Neifert, Timken Roller Bearing Co.

Chip Formation, Friction, and High Quality Machined Surfaces, by Hans Ernst and M. E. Merchant, Cincinnati Milling Machine Co.

Observations on the Tarnishing of Stainless Steels on Heating in Vacuo, by V. C. F. Holm, National Bureau of Standards.

The Tracer Method of Measuring Surface Irregularities, by E. J. Abbott, Physicists Research Co.

The following papers were presented by title:

Alloys of Manganese and Copper: Electrical Resistance, by R. S. Dean and C. T. Anderson, Bureau of Mines, U. S. Dept. of Interior.

Alloys of Manganese and Copper: Hardening by Cold Work and Heat Treatment, by R. S. Dean and C. T. Anderson, Bureau of Mines, U. S. Dept. of Interior.

Alloys of Manganese-Copper and Nickel; Hardening in the Pseudo Binary System Cu-Mn-Ni, by R. S. Dean and C. T. Anderson, Bureau of Mines, U. S. Dept. of Interior.

PLANT INSPECTION

The following companies very kindly opened their plants for the inspection of members and guests of the Society.

Tuesday, October 22

Lakeside Steel Improvement Co.

White Motor Co.

Tinnerman Products, Inc.

Wednesday, October 23

American Gas Association Testing Laboratories

General Electric Institute, Nela Park

Republic Steel Corp.

Thursday, October 24

Lamson & Sessions Co.

Otis Steel Co.

Ohio Crankshaft Co.

Friday, October 25

Eaton Manufacturing Co.

National Acme Co.

Cleveland Tractor Co.

ANNUAL ADDRESS OF THE PRESIDENT

Twenty-Second Annual Convention, Cleveland, October 23, 1940

JAMES P. GILL, *President*

YOUR Society has again enjoyed one of its usual successful years of progress and growth during the term of office of your retiring President but I assure you little of this success can be attributed to him. To the members of this Society, to their Chapter officers, to the National Committees, to the Board of Trustees, to the Secretary and his splendid co-operating staff must be given the credit of the continued progress and widening usefulness of the Society. To each of you the President welcomes this opportunity to thank you for your splendid help and co-operation during the past year.

The Board of Trustees held four meetings during the year, November 17, 1939, April 12, 1940, August 16, 1940 and October 22, 1940. A detailed report of each meeting was published in the issue of REVIEW that followed.

While the new headquarters were purchased in March, 1939, we did not receive possession until November 1, 1939. Several months were required for the few necessary changes to adapt the building for the Society's use. Consequently, we were not established in our new home until the end of February of this year. All of the members of the Society who have had an opportunity to visit the headquarters are pleased and proud of the manner in which the building is adapted for our requirements. Past President Woodside reported the purchase price at \$45,000.00 and the Society found it necessary to spend about \$8000.00 additional to adapt the building for headquarters purposes. The Board of Trustees have determined that the property is to have a complete depreciation at the end of the fifteen-year period.

The Past President's medal was awarded to Dr. George B. Waterhouse at the annual banquet at Chicago on October 26, 1939. The Henry Marion Howe medal, awarded to the paper of highest merit published in TRANSACTIONS during the American Society for Metals year ending in August 1938, was awarded at the annual banquet in Chicago on October 26, 1939, to C. S. Barrett, G. Ansel and R. F. Mehl for their paper entitled "Slip, Twinning and Cleavage in Iron and Silicon Ferrite" published in the September, 1937, issue of the TRANSACTIONS. After a most careful study the Committee on award of the Henry Marion Howe medal have reported

no recommendations of award for the American Society for Metals year ending in August of 1939.

The Campbell Memorial Lecture, established in 1925, was presented in 1939 by E. S. Davenport, U. S. Steel Corporation, Res. Lab., entitled "Isothermal Transformation in Steels", and was published in the December, 1939, issue of *TRANSACTIONS*.

The Sauveur Achievement Award, established in 1934, is awarded by the Past Presidents of the Society acting as a committee. The award for 1939 was presented at the annual banquet in Chicago to Stanley P. Rockwell in recognition of his pioneering work in metallurgical science in the development of the Rockwell Hardness Testing Machine and the Rockwell Dilatometer. It is with deep regret that mention must be made of Mr. Rockwell's accidental death during the past summer.

Your President is pleased to announce that the Board of Trustees have approved an appropriation of \$3000.00 for co-operative research on visual education between the American Society for Metals and the Ohio State University Research Foundation. The primary purpose of this research is:

1. The survey of all available visual materials in the field of metallurgy and note any portions of these materials which may be suitable for our use.
2. To secure the judgment of a limited number of teachers and specialists in this field in order to discover those phases of metallurgical education which they believe lend themselves readily to visual presentation.
3. A completion of the silent film study entitled "Metallic Crystals" which was started by members of the Educational Committee and is now nearly half finished.
4. The consideration of other subjects and preparation of either movies of film-slide-sound records of those that appear most promising.

The Society is under deep obligation to Dr. Maxwell Gensamer for his series of five lectures on the subject of "Behavior of Metals Under Stress" which he is presenting during this Metal Congress and if Dr. Gensamer will be kind enough to come forward to the platform I wish to present to him on behalf of the Society this certificate of appreciation for his lectures. The Society likewise is under deep obligation to Dr. A. A. Bates, Prof. W. J. Conley and Mr. R. G. Roshong for their series of three lectures on the "Quenching of

Steel" and if Dr. Bates, Professor Conley and Mr. Roshong will come forward I also wish to present them with a certificate of appreciation for their activities during this week. These educational lectures have become outstanding features and it was a pleasure of your present President to have inaugurated this series of lectures at the New York Convention in 1934.

Our country is now engaged in building a system of national defense requiring the resources of much of our manufacturing facilities. As individuals or as groups, the members of American Society for Metals have expressed a willingness to help in this defense program to the best of their ability. A telegram was addressed to Mr. Stettinius as follows:

"With the authority of the Board of Trustees and the eleven thousand members of the American Society for Metals I offer the committee for national defense their full co-operation and desire to serve as individuals or on committees in any and all ways in which their training and experience in the manufacture, fabrication, treatment or inspection of metals may be required. The metallurgical profession in the American Society for Metals is subject to your call."

—James P. Gill, *President*

Mr. Stettinius replied:

"Thank you for your wire of June 14th. I deeply appreciate the patriotic offer of the services of the American Society for Metals, which I am turning over to our consultant on metals."—E. R. Stettinius, Jr.

Undoubtedly there are many ways in which the Society members, individually and as groups, can help in the present emergency. The Philadelphia Chapter has forcibly demonstrated one way in which the Society can prove a real benefit at the present time. The Philadelphia Chapter organized a series of educational lectures in metallurgical inspection purposely adapted for Government inspectors in the field of metals. No words are needed to show the success of this series of educational lectures other than to mention that the attendance has been from 700 to 800. That these efforts were appreciated is reflected by a letter of commendation written to the Philadelphia Chapter by the Ordnance Department in Washington.

Other ways in which the Society and its members can prove of benefit in the national defense program is by the assistance of committees in the work of any metallurgical problem which the Government or industry may have; by meetings of members of the Society

with the personnel of the Ordnance Department such as the meeting which will occur at this Convention on Friday when Major General Wesson, Chief of Ordnance, will confer with those in attendance at this Convention in an endeavor to acquaint industry with the requirements of the Ordnance Department and to secure a further co-operation of industry in the manufacture of items for national defense.

The Society relishes the opportunity which has been afforded to it to serve as the medium for bringing these two important groups together for such conferences.

I am sorry to announce the death of one of our famed honorary members, Sir Robert Hadfield, who died several months ago at the age of 81.

Your President has in the last few years addressed every Chapter of the Society and during his term as President, by traveling some 30,000 miles, was able to visit with and talk to 44 of the 52 Chapters. The President has enjoyed the company of the Secretary on these Chapter visits. Such visits have provided excellent opportunity to discuss the work of the Society, its deficiencies and how improvements could be instituted. Three new Chapters have been established: The Rocky Mountain Chapter with the focal point of Denver, Colorado; the Southern Chapter in the district of Birmingham, Alabama, and the Northwest Pennsylvania Chapter in the Meadville-Erie district. These Chapters have given evidence of considerable strength and stability and it is with real pleasure that we welcome them into the Society.

Our Society has become big and it has become difficult, if not impossible, for the President to have intimate knowledge of the activities of the individual Chapters. This makes it difficult, and to some extent embarrassing, for the President to make the annual award of the President's Bell. Upon what features of a Chapter's activities should the Bell be awarded? By the activities of the membership committee; by that of the educational committee; by the percentage of the Chapter membership meeting attendance; by the quality of the programs, etc.? The enthusiasm, hard work, the ideas of all of the Chapters which the President visited was excellent. We simply do not have any mediocre or poor Chapters, but the Board has told your President that he must award a Bell but that after this year they would give detailed study as to whether or not a system could be evolved that would make the awarding of this Bell one of

easier decision. This year the President takes pleasure in awarding the Bell to the Texas Chapter so I ask the Chairman, Wade Hampden, to please come forward to receive the Bell.

The President takes this opportunity in behalf of the American Society for Metals to acknowledge with sincere thanks the cooperation of the Societies which have associated themselves in making this Congress a great success: The American Welding Society; The Wire Association; and Iron and Steel Division and the Institute of Metals Division of the American Institute of Mining and Metallurgical Engineers.

REPORT OF THE TREASURER

KENT R. VAN HORN, *Treasurer*

The fiscal year, ending on August 31, 1940, has been another successful year in the history of the Society, and it is a satisfaction to your Treasurer to submit this report.

Our cash balance on August 31, 1940, was \$120,708.22 against last year's \$100,427.42. We are, therefore, in a very liquid position and can take advantage of any opportunities or a drastically depressed market.

Security holdings of the Society increased the past year from \$182,147.54 to \$199,744.34, or 9.6 per cent. Several investments were called for redemption; but additional purchases replaced these holdings, and surplus funds permitted additional investments to be made. Total assets increased 13.6 per cent; deferred income and expense on the present Convention will, however, when the show accounts are complete, reduce the increase to approximately 10 per cent.

Accounts Receivable decreased in total (from \$12,875.27 to \$8477.52), showing the results of several collections on delinquent accounts and a very healthy condition of receivables, which have an average collection period of less than 30 days. This is the lowest position of Accounts Receivable since 1934, when our operations were at a much lower rate.

Inventories have increased approximately \$7400.00 due to the publication of several new books and additional paper stock purchased at the end of the fiscal year.

Office furniture and equipment account has increased approximately \$3500.00 due to additional floor coverings and equipment needed for the new headquarters.

The Investment in the headquarters building was increased by \$8000.00 for necessary alterations and improvements, but amortization of this investment over a fifteen year period has been authorized by the Board of Trustees.

The Finance Committee is in constant touch with the A.S.M. fiscal agents, the Cleveland Trust Company; and a close contact with the investment market is maintained. In fact, the Trustees have suggested frequent meetings of the Finance Committee to consider our investments during these uncertain times. In addition, because of the residence of both your Secretary and Treasurer in Cleveland, it has been possible to hold frequent counsels with Mr. W. W. Horner, Vice-President in charge of investments of the Cleveland Trust Co., and Dr. Zay Jeffries, senior member of the finance committee. Addition of top grade stocks to the portfolio during the present year has increased the hedge against inflation, will help to maintain dollar value, and will increase diversification.

The investment portfolio now shows:

		Per Cent
Real Estate (Headquarters)	\$51,611.00	20.5
U. S. Gov't. Bonds	7,500.00	3.0
Railroad Bonds	35,930.00	14.3
Public Utility Bonds	45,750.00	18.2
Industrial & Misc. Bonds	15,970.00	6.0
Common Stocks	94,595.00	38.0

Including the cash on hand, common stocks now represent 25.4 per cent of the investments. At a recent meeting, the Board of Trustees voted to increase this percentage when and if opportunities appeared. As of August 31, 1940, the total cost of present investments (not including Headquarters) was \$199,744.34, and the market value was \$155,540.00. From 1921 to August 31, 1940, the investments have yielded \$102,693.63 and the sale of securities during this period has resulted in a trading profit of \$6366.61.

The present market values of securities	\$155,540.00
and Headquarters	51,611.60
Plus cash on hand	120,708.22
	<hr/>
	\$327,859.82

This total exceeds by only \$32,000.00 the total operating cost of the Society and its activities for 1939-40, \$295,167.86. The in-

terest at 3 per cent that could be realized from this sum of \$327,859.82 is, therefore, approximately only $3\frac{1}{2}$ per cent of the total operating expense (Society and Convention). The Finance Committee believes it is advisable to strive to maintain a substantial reserve.

The statement of income and expense for the year ending August 31, 1940, is as follows in total:

Income	\$332,035.17
Expense	295,167.86
Excess	\$36,867.31

This gain is exceeded only by the years 1927, 1937, and 1939 in our Society's records. Compared to \$47,811.00 for 1939, it would seem that the decrease is due to a smaller revenue from the Chicago Convention compared to Detroit, and particularly to the extended scope of our Society's activities and increased benefits to a larger membership.

Income from METAL PROGRESS increased approximately 15 per cent this year over the prior year from \$104,396.86 to \$119,433.00 and expense increased 9 per cent from 99,518.86 to 108,302.06 increasing the net gain from \$ 4,878.00 to \$ 11,130.94

This fiscal year METAL PROGRESS has shown a gain appreciably exceeding any of the previous years. Now METAL PROGRESS contributes about one-third of the annual income.

TRANSACTIONS and THE REVIEW are, of course, not self-sustaining and cost the Society a net sum of approximately \$16,500.00 and \$4600.00.

Miscellaneous items such as reprints, books purchased for resale, pins and buttons, automobile emblems, etc., balance out. Books published, however, add a net gain of approximately \$5800.00. This figure, like 1939, shows a substantial reduction from 1937 and 1938. This is attributed to the recent policy of offering publications to the membership at reduced prices on publication and at certain intervals.

The 1939 edition of the METALS HANDBOOK continues as a best seller, and sales of \$8500.00 for the past year have been added to sales previously reported of \$6200.00.

The 1939 Convention and Exposition, as usual, was an outstanding event in the metal industry and was successful, not only financially, but artistically, as a selling and advertising medium for our exhibitors. While the financial gain was only half of the previous year's show, the added value of the expanded technical and educational programs more than compensated for this.

CONDENSED AUDITED BALANCE SHEET

As of August 31, 1940

ASSETS

Cash	\$120,591.07
Securities	170,689.12
Accounts receivable	10,546.55
Inventories	36,862.34
Other assets	4,894.02
Real estate	51,611.60
Office furniture, fixtures and equipment	12,156.31
Deferred charges	22,844.17
	<u>\$430,195.18</u>

LIABILITIES AND SURPLUS

Accounts payable	\$ 5,600.23
Accrued taxes	1,606.19
Reserves	110,000.00
Deferred income	67,414.75
Surplus	245,574.01
	<u>\$430,195.18</u>

INCOME AND EXPENSE STATEMENT

Year Ended August 31, 1940

INCOME

METAL PROGRESS—monthly publication:

Advertising	\$113,463.28	
Subscriptions	3,579.95	
Advertisers service	2,190.47	
Sales	199.30	\$119,433.00

MEMBERSHIPS:

Membership dues	\$115,502.54	
Exhibitor memberships	5,250.00	\$120,752.54
Less apportionment of dues to chapters	46,248.54	74,504.00

1939 CONVENTION—CHICAGO:

Space rentals	\$ 75,323.50	
Other income	14,128.95	89,452.45

Metals Handbook—sales	8,526.54
Books published and purchased for resale	20,746.87
Interest and dividends	6,220.02
TRANSACTIONS—bound copies, subscriptions, etc.	3,865.22
General reprints	3,631.75
Discount earned	2,201.80
Membership certificates	668.80
Chapters—technical service	654.85
Profit on disposal of securities	512.34
Rental income—7301 Euclid Avenue	425.44
THE REVIEW—advertising and subscriptions	37.14
Sundry	1,154.95

TOTAL INCOME \$332,035.17

EXPENSES

As shown by schedule:

Metal Progress—monthly publication	\$108,302.06
1939 Convention—Chicago	74,676.27
Metals Handbook	8,133.34
Books published and purchased for resale	15,425.08
TRANSACTIONS—bound copies, subscriptions, etc.	19,546.53
General reprints	3,230.57
THE REVIEW	4,669.33
General expenses	14,146.58
Headquarters	5,732.67
Secretary's office	13,671.15
Membership	10,780.14
Accounting department	7,101.87
Expense of trustees	3,501.10
Expense of president	3,234.56
National committees	1,446.99
	<u>\$293,598.24</u>

Lectures, medals, etc.	1,025.70
Books for library	162.20
Sundry	381.72

TOTAL EXPENSES \$295,167.86

NET INCOME BEFORE SURPLUS CHARGE \$ 36,867.31

The revenue from membership dues reached a new high this last fiscal year.

Gross total for 1940	\$115,502.00
1939	108,000.00
1938	103,600.00
1937	105,600.00

In 1940 the income from membership represented only 34.8 per cent of the total income, or 39.7 per cent of the cost of operating the Society. \$46,248.00 of this gross was returned to the chapters.

The chapters now report that they have assets, in addition to those of the National Office, totaling \$80,500.00, an addition of \$7900.00 to the net worth reported a year ago. This reserve substantially insures the continued activity and progress of the chapters.

Your Treasurer wants to gratefully acknowledge the active co-operation and assistance of the Finance Committee, the Board of Trustees, President Gill, Secretary Eisenman, and Mr. C. W. Ohlson.

ANNUAL REPORT OF THE SECRETARY

WILLIAM H. EISENMAN, *Secretary*

The American Society for Metals on October 1, 1940 had a total membership of 11,122. Of this number 9809 or 88.19 per cent were the member classification; 684 or 6.14 per cent were sustaining members; while 604 or 5.43 per cent were juniors. There are 23 honorary and founder members.

On October 1, 1939 the Society had a total membership of 10,220. The membership this year showed a gain of 8.92 per cent.

Publication Committee

The Publication Committee for the year 1939-40 was made up of the following: L. W. Kempf, Chairman; L. S. Bergen, A. L. Boegehold, W. H. Bassett, Jr., J. L. Burns, M. Gensamer, R. H. Hobrock, J. J. Kanter, B. L. McCarthy, M. J. R. Morris, J. F. Oesterle, W. H. Swanger, Sam Tour, J. P. Walsted, and A. W. Winston.

Throughout the year and up to the present time the Committee has reviewed and approved 60 papers, rejected 10, and is in the process of reviewing 3 at the present time, and 5 have been withdrawn by the authors.

The last meeting of the Publication Committee was held in

National headquarters office in Cleveland on August 9 and 10, 1940. This meeting was concerned with the arrangements of the technical program for this Convention. The selection of the subject for the Symposium for the 1941 Convention will be on "Controlled Atmospheres."

Transactions

There has been published, since the last annual meeting of the Society, four quarterly issues of TRANSACTIONS, i.e., December 1939, March, June and September 1940. The number of pages published in these four issues totals 1078. There were 41 articles printed together with the discussions. A total of 45,523 TRANSACTIONS has been distributed to the membership.

The TRANSACTIONS published since December 1939 contain papers presented at the 1939 Convention together with others received during the year.

Metal Progress

Statistics for the last four fiscal years (appended) show that 1940 has been the best in volume of advertising (in fact, the best year in METAL PROGRESS' history). Net paid space at 915 pages is almost a hundred more than carried in 1939. For this satisfactory showing Messrs. Ford, Wells and Byrne deserve congratulations. Editorial pages printed, at 662½, are some 50 more than an average year, due to the large anniversary issue published in March. Average circulation, in keeping with the enlarged membership, reached an all-time high. Notwithstanding these larger expenses, the balance sheet shows the best figure for surplus of METAL PROGRESS income over expenditure of any year during the ten of its existence. The good prospects for next fiscal year may be gaged by the October Annual Reference Issue, recently delivered to you.

Fiscal Year	Net Paid Advertising Pages	Editorial Pages
1937	867½	633
1938	877½	613½
1939	816	612
1940	915	662½

Such statistics are unable to measure the importance of this magazine as a Society enterprise. Many expressions of appreciation from members are received by the national office. For the third year

in succession METAL PROGRESS has received an "Award of Merit for Editorial Excellence" in the competition conducted by *Industrial Marketing* and sponsored by the American Industrial Advertisers' Association. It also gives me pleasure to quote the following from the minutes of the June 13 meeting of the Editorial Advisory Board for METAL PROGRESS:

"Mr. Phillips moved (seconded by Mr. Crowe) that the Editorial Advisory Board congratulate the Editor, Ernest E. Thum, most heartily on the very successful Anniversary Issue for March 1940, and that this action be recorded in the minutes. The motion was duly put and carried."

In conveying the thanks of the Society to the hundreds of members who have co-operated with us, both in securing editorial matter and advertising patronage, I wish to remind all other members that METAL PROGRESS is their joint property, that it is unapproached in appearance and quality, that it has the cream of metallurgical circulation. It should therefore be their *first* thought when they learn of some desirable and practical article that is ready for publication.

Preprints

Forty-eight papers were presented before the 21st Annual Convention held in Chicago last year. For this year's Convention 58 papers have been scheduled for presentation, 55 of which have been preprinted and distributed to those members requesting them. The total number of *pages* for 1940 is 1260. A total of 43,777 preprints have been distributed free to the membership to date.

Books

Since the last Convention, five books have been published and made available to members and others, i.e., "*Pyrometry of Solids and Surfaces*" by Dr. R. B. Sosman; "*The U. S. Patent System*" by John A. Dienner; "*Practical Metallurgy*" by Dr. George Sachs and Dr. Kent R. Van Horn; "*Modern Steels*" which consists of a series of lectures presented before the Pittsburgh Chapter; and "*Age Hardening*" which consists of the papers which were presented in the 1939 Symposium. The 5-lecture series on "*The Strength of Metals Under Combined Stresses*" which is being presented at the 1940 Convention by Dr. Maxwell Gensamer will be prepared in book form shortly after the Convention. The 3-lecture series by Dr. A. A. Bates, Prof.

W. J. Conley and R. G. Roshong on the subject of Quenching will also be prepared in book form shortly after the Convention.

The Review

During the past year THE REVIEW has been published monthly with the exception of July and September. Of the 10 issues which appeared from November 1939 to October 1940 all contained 8 pages with the exception of the November 1939 and the October 1940 issues which carried 12 pages each.

The October issue was devoted almost entirely to the National Metal Congress and Exposition and had a circulation of 26,000 among the members of the co-operating societies in the Congress and other technical organizations.

Of the total space used in the 10 issues during the past year, 3500 column-inches or 79.5 per cent was devoted to editorial matter, and 900 column-inches or 20.5 per cent to advertising. Reports of chapter meetings and educational activities used 1919 column-inches or 43.6 per cent of the total space.

Metals Handbook Committee

The personnel of METALS HANDBOOK Committee is as follows:

R. S. Archer, Chairman, Republic Steel Corp.
E. L. Bartholomew, United Shoe Machinery Corp.
A. D. Beeken, Jr., Vulcan Crucible Steel Co.
W. P. Eddy, Jr., General Motors Truck & Coach Co.
J. H. Hall (Resigned September 1940)
H. L. Maxwell, E. I. duPont de Nemours & Co.
R. F. Mehl, Carnegie Institute of Technology
C. W. Obert, Union Carbide & Carbon Corp.
H. B. Pulsifer, American Steel & Wire Co.
A. O. Schaefer, Midvale Co.
S. C. Spalding, American Brass Co.
A. P. Spooner, Bethlehem Steel Co.
Lyall Zickrick, General Electric Co.

During the last year, METALS HANDBOOK Committee has concentrated its efforts on preparing a program and formulating policies for the preparation of the next edition of METALS HANDBOOK. The committee has devoted much time during the last year in studying the present issue of the HANDBOOK to determine what sections can be condensed so that the next edition may be contained in one volume, to determine what articles might be eliminated, and to decide what new articles should be added. To answer the questions, the committee has studied every article in METALS HANDBOOK so that definite

recommendations can be made to authors and subcommittees for condensing some articles to a certain number of pages and for controlling the number of references incorporated in these articles. With this plan and by limiting future authors and subcommittees to a certain number of pages for new articles, the committee feels quite confident that the next issue of the HANDBOOK can be prepared in one volume.

METALS HANDBOOK Committee has had two regular meetings during the last year. Both meetings were held in the National Office of the Society; the first on January 18th, and the second one, which was a two-day meeting, on June 14th and 15th.

As a result of these two METALS HANDBOOK Committee meetings, 50 new subjects were selected for incorporating in the next edition of METALS HANDBOOK, 15 subcommittees were organized and 81 authors selected to prepare these articles.

The appointments for these subcommittees and authors are now in process of being made so all acceptances are not in, but the indications are very favorable that the committee will again receive the same splendid co-operation and assistance from the membership in the preparation of the next edition of METALS HANDBOOK.

The 1939 edition of METALS HANDBOOK has now been available 18 months. During this time 13,020 books have been issued. 1696 of these have been sold, leaving 11,324 that have been issued to members of the Society in exchange for their 1936 edition. This leaves 1630 members of the Society who have still not taken advantage of the free exchange arrangement available to them for the 1939 METALS HANDBOOK. The Society still has 6980 books in stock which will take care of the requirements for some time so there are no plans as yet for issuing a new edition.

The International Acetylene Association is again going to co-operate with METALS HANDBOOK Committee by preparing the sections on Welding and Cutting of Metals. We are pleased with this continued co-operation because the International Acetylene Association Committee and especially the chairman, Mr. C. W. Obert, have contributed valuable information for the HANDBOOK on these subjects. The personnel of the International Acetylene Committee was as follows:

C. W. Obert, Chairman, Union Carbide & Carbon Corp.
C. H. Jennings, Westinghouse Electric & Mfg. Co.
H. L. Maxwell, E. I. duPont de Nemours Co.
L. D. Meeker, General Electric Co.

E. V. David, Air Reduction Sales
E. W. P. Smith, Lincoln Electric Co.
W. L. Warner, Watertown Arsenal
A. C. Wikoff, Union Carbide & Carbon Corp.

The Nonferrous Data Sheet Committee of the Institute of Metals Division of the A.I.M.E. held a meeting in New York on April 25th. At that meeting the committee prepared a program for improving and enlarging the Nonferrous Section of the HANDBOOK. The recommendations have been submitted to METALS HANDBOOK Committee and the work of organizing subcommittees and selecting authors is now in process. The Nonferrous Data Sheet Committee has contributed a very important and valuable section of METALS HANDBOOK and we are pleased with this continued co-operation in the preparation and revision of the Nonferrous Section of the next edition of METALS HANDBOOK. The personnel of the present Nonferrous Data Sheet Committee is as follows:

Lyall Zickrick, Chairman, General Electric Co.
W. H. Bassett, Jr., Anaconda Wire and Cable Co.
W. E. Remmers, Electro-Metallurgical Sales Corp.
E. E. Schumacher, Bell Telephone Laboratories
S. Skowronski, Raritan Copper Works
T. A. Wright, Lucius Pitkin, Inc.

Since issuing the 1939 edition of METALS HANDBOOK, many comments, suggestions, and recommendations have been made by members of the Society. All of these suggestions have been studied by METALS HANDBOOK Committee and will result in many improvements in future issues of the HANDBOOK. They have been most helpful to the committee in formulating plans and preparing a program for future HANDBOOK work.

A.S.M. Educational Committee

The Educational Committee for the year 1940 was made up of the following: R. L. Kenyon, Chairman; E. C. Bain, A. A. Bates, W. J. Conley, H. P. Croft, H. C. Knerr, W. M. Saunders, Jr., and Ray T. Bayless, Secretary.

The Committee has held two meetings since the last Convention: one on November 9, 1939, and the second on May 28, 1940.

The meeting in November was called chiefly for the purpose of conferring with Dr. Maxwell Gensamer who had been selected by the Committee to give a five-lecture series on the subject of "Strength of Metals Under Combined Stresses" which is being presented at

this Convention. This gave both the author and the Committee an opportunity of reviewing the proposed context of the lectures and each was able to offer suggestions to the other in the preparation of this manuscript.

This meeting also concerned itself with the consideration of the three-lecture series for this Cleveland Convention on the subject of "Quenching of Metals" and definite arrangements were made with the three authors, i.e., A. A. Bates, W. J. Conley, and R. G. Roshong, for the presentation of this series of lectures.

Chairman Kenyon reported on the progress that had been made in the preparation of the moving picture entitled "Metallic Crystals" and was confident that much progress would be made in the very near future in completing this film.

Considerable discussion was held as to possible subjects for the 1941 lecture series and a sizable list of subjects was brought up for consideration.

The meeting of May 28th was held in headquarters office at Cleveland and concerned itself chiefly with the work that was being carried out by the Committee in the preparation of the moving picture film entitled "Metallic Crystals." Chairman Kenyon presented sample "shots" of portions of the scenario and demonstrated a sound slide projector that might be of considerable value as equipment for visual educational purposes.

As a result of this demonstration the Committee went on record as favorably inclined toward this type of equipment for educational lectures.

Dean C. E. MacQuigg of the College of Engineering, Ohio State University, attended this meeting upon invitation of the Chairman and discussed the many phases of the work that Ohio State University is doing on visual education, and outlined how the University could assist the A.S.M. in developing methods of visual education.

As a result of this conference with Dean MacQuigg it was requested that the Chairman and Dean MacQuigg proceed with further investigations as to how precisely Ohio State University and the A.S.M. could work together in developing or creating satisfactory methods of visual education for the use of the chapters of the National Society.

As a result of further investigation by these two men the Board of Trustees of the A.S.M. has approved an arrangement for co-opera-

tive investigation with Ohio State University Research Foundation and the Society to develop new methods of visual education.

In the research project now planned, for which the A.S.M. will bear the expense (approximately \$3000), the Foundation will furnish available facilities of space and laboratory equipment. The services of a graduate fellow in metallurgy, a graduate fellow in photography, and a graduate assistant in education will be provided, and the special equipment necessary.

The agreement has been entered into for the period from October 1, 1940 to June 30, 1941. The work will have the following objectives:

1. Survey of all available visual materials in the field of metallurgy and note any portions of these materials which may be suitable for our use.
2. Secure the judgment of a limited number of teachers and specialists in this field in order to discover those phases of metallurgical education which they believe lend themselves readily to visual presentation.
3. Completion of the silent film study entitled "Metallic Crystals", which was started by members of the Educational Committee and is now nearly half finished.
4. Consideration of other subjects and preparation of either movies or film-slide-sound records of those that appear most promising.

The work will be under the supervision of Dr. A. R. Olpin, research director of the Ohio State University Research Foundation, and Reid L. Kenyon of the American Rolling Mill Co., chairman of the A.S.M. Educational Committee.

Employment Service

The change in the A.S.M. employment service which was inaugurated October 16, 1939, under the direction of the Technical Placement Service of Cleveland and guidance of T. G. Protheroe, its President, and amplified by the services of 32 offices through the United States, has been functioning actively throughout the year. The results that have been obtained from this service thus far are shown in the following tabulation which covers the period of October 16, 1939 to October 16, 1940:

Number of applications received	188
Number employed	114
Number personally interviewed at T.P.S.	30
Number diff. chapters from which applications come	35
A.S.M. members referred to positions from master file	148
Calls received at T.P.S. of A.S.M. origin	42
A.S.M. members referred on above calls	99
A.S.M. members referred on calls not of A.S.M. origin	57
Total number of referrals	176
Placements	9
Pending	15
Calls received through Assoc. Offices of A.S.M. origin	9
Calls received from the A.S.M. main office	32

Thus, out of a total of 188 A.S.M. members listed, there have been 176 referrals to positions. Naturally, some of the members have been referred to more than one position. Of the above 176 referrals, we know of only 27 of these men who have had personal interviews. Of these, 9 obtained positions—about 30 per cent, which is a good percentage. The balance of the referrals have all been through the mail.

Western Metal Exposition

Your Board of Trustees has decided to hold the next Western Metal Congress and Exposition in Los Angeles the week of May 19, 1941. This will be the fourth congress and exposition to be held on the coast.

The first was in 1929, the second in 1931, the third in 1938. They were all outstandingly successful.

Many national technical societies and associations having branches and sections in the West have been invited, as in previous congresses, to participate in the arrangements of the technical program which will have many interesting and unique features.

The Western Metal Exposition will show a complete cycle of the metal industry.

The motivating factor in holding the fourth Western Metal Congress and Exposition is the sincere desire and purpose of the Board of Trustees and the Society to be of the same service to the metal industries of the West as the National Metal Exposition and Congress are to the industries of the East.

National Metal Exposition

This present congress and exposition eclipses in magnitude and interest those of previous years. The congress has attracted a large

number of outstanding papers and the national situation has made the authors' contributions to the science of metals of outstanding interest.

The National Metal Exposition with its theme, "New Aids to Production," has gathered together 330 exhibitors occupying 103,000 square feet of exhibit space. This is the largest exposition in the 22-year history of the Metal Show and represents a growth of 29 per cent over the 1939 show in Chicago.

After careful consideration, and following the custom of having every fourth show in the East, the Board of Trustees decided that the 1941 congress and exposition would be held in the city of Philadelphia.

Sauveur Award Made by President Gill

A few years ago, the Sauveur Achievement Award was instituted and awarded at the annual banquet. This award is made by a vote of the Past Presidents of the Society. This year the award has been given to Mr. A. W. Machlet of the American Gas Furnace Company and as Mr. Machlet cannot be present at our annual banquet on Thursday evening, I am taking this opportunity to tell you briefly of some of the outstanding accomplishments of Mr. Machlet which have been the basis for this award.

"Mr. Adolph W. Machlet is the second of three sons of cultured parents, now deceased, of German extraction. The three sons, as was their father, George W. Machlet, were all of inventive turn of mind and all three, upon leaving school, became associated with the American Gas Furnace Company, in which firm their entire business careers were passed. This business was established in Elizabeth, N. J., in the 1870's, jointly by Major Edward P. Reichhelm, a Civil War Veteran, and George W. Machlet, upon the latter's idea of combustion which was the reverse of Professor Bunsen's then generally known system.

"As is well known, the Bunsen system is dependent upon the gas stream entraining the needed volumes of air for combustion, whereas in George W. Machlet's system of combustion the air under pressure entrains the requisite volumes of gas, and this latter system has since been almost universally adopted for gas-fired industrial furnaces.

"George W. Machlet's three sons were George F., now deceased; Adolph W., the subject of these notes; and Fritz W. Machlet.

"The many inventions of processes and apparati by Adolph W. Machlet have very markedly accelerated progress and development in the metallurgical field and have to a considerable extent been of public benefit because of greater economy, better performance and increased safety in various mechanisms in the production of which these inventions have had a part.

"His basic work in bringing out the *Nitriding Process* blazed a trail which daily is increasing in its scope of usefulness in the treatment of steels for wear and corrosion resistance. The application of this process is being employed in ever widening scope, notwithstanding that his basic Nitriding patent expired several years ago.

"To him must be credited the so-called '*Carbonia*' Process, that is, the production of a surface finish superior in appearance and durability to processes theretofore in use. This process is employed by Government Armories and practically all manufacturers of small arms in imparting to their products the pleasing black or blue-black surface finish. The '*Carbonia*' Process is carried out in rotary machines. The finish which it displaced was obtained by long and tedious hand work. This process is also extensively used on many other ferrous articles such as typewriter parts, washers, screws, metal fastenings, etc.

"He also invented the process known as '*Ni-Carbing*' which, as its name implies, is a combination of nitriding and carburizing, applicable successfully to ferrous articles of a wider range of steels and iron than is practical in the case of Nitriding alone.

"One of the most universally used heat treating processes on steel articles is the one known as case-hardening. Mr. Machlet's basic work in inventing and developing a direct *Gas Carburizing Process and Machine* has very measurably cheapened and bettered the case-hardening process over any method of practicing it theretofore. His process produces the desired results of better quality at lower cost and more expeditiously, and its adoption improves greatly the working conditions for the operatives in heat treating departments.

"One of his latest developments is a superior method for the clean hardening of carbon and high speed steels. By this method, the steel is protected against oxidation or other detrimental attack from start to finish of the process."

Mr. Machlet's associate of many years, Mr. P. C. Osterman, is present this morning and I have asked him to receive this plaque and certificate and deliver them to Mr. Machlet as the Society's recognition of his outstanding accomplishments.

Mr. Osterman received the certificate and plaque and responded by reading the following letter from Mr. Machlet.

"My hearty greetings to you on this, another of your important yearly events. May this convention be as successful as those which have preceded it. Perhaps this wish is superfluous, for it seems to be a foregone conclusion that whatever is contemplated under the leadership of your capable Secretary, Mr. Eisenman, with the co-operation of your able committees, must succeed.

"For the honor which you have just conferred upon me, namely the Sauveur Achievement Award in the form of a plaque and certificate, please accept my heartfelt thanks.

"I understand this honor is presented upon recommendation of a committee composed of your Past Presidents, some of whom I have the pleasure of knowing personally, the others I know by their reputations. All of these gentlemen have distinguished themselves as engineers, metallurgists or scientists; therefore, the greatest satisfaction comes to me from the belief that this honor is an ex-

pression by the Society through this committee of Masters, that my efforts have been of value to others in the metallurgical profession.

"Incidentally, as my seventy-sixth birthday is on October 24th, I shall cherish this honor as a memento of that anniversary.

"I regret exceedingly that I cannot be with you personally, as I find it rather arduous to travel, and trust, therefore, that you will kindly bear with my request that you permit Mr. P. C. Osterman, who has been associated with me almost forty years, to act as proxy for me on his important occasion.

"Again, I thank you most sincerely.

Respectfully yours,

ADOLPH W. MACHLET"

Complete List of Mr. Machlet's Patents

Following is what is believed to be a complete list of A. W. Machlet's patents with descriptive details on some of them.

11,069, May 11, 1906, England. Improvements in case hardening of steel and iron articles.

191,394, May 23, 1906, Germany. Verfahren und Oven zum gleichmaessigen Zementieren von Massenartikeln aus Eisen und Stahl auf eine vorbestimmte Tiefe mit Hilfe von kohlenwasserstoffhaltigem Gas.

822,460, June 5, 1906; U. S. A. Casehardening Apparatus. Applied for February 4, 1905, patented June 5, 1906. This is the first patent issued to A. W. Machlet on *Gas Carburizing Machines*. In this machine the work was discharged from a port in the retort through the heating chamber, i.e., the discharge port of the retort was in the heating chamber.

366,245, July 24, 1906, France. Procédé et appareil pour la cementation des objets en fer et en acier.

847,588, March 19, 1907, U. S. A. Casehardening. Applied for April 13, 1905, patented March 19, 1907. This is a *Gas Carburizing Process* (companion patent to 822,460). As to the kind of carburizing gas, quoting from the patent: "The carburizing gas may be a pure hydrocarbon, although I have found in practice that good results are obtained by the use of ordinary city gas, such as generally supplied for illuminating and heating in the city of Elizabeth, N. J. Good results can also be obtained from coal gas and from oil gas and also from carbureted hydrogen gas (or water gas) resulting from the passing of steam through a mass of incandescent carbon and subsequent admixture of hydrocarbons or other enriching substances."

As to the pressure of the carburizing gas, quoting from the patent: "The gas which is used for cementation may be supplied under pressure in any suitable way; but for the sake of illustration I have shown a gas receiver or tank 40, which contains gas preferably at a pressure of one hundred pounds to the square inch, although a much lower or even higher pressure may be used, or in some cases the pressure may be just a little higher than atmospheric pressure. In other words, the pressure may vary from just sufficient to charge the retort to the highest that may be found practicable."

857,292, June 18, 1907, U. S. A. Continuous Pusher Type Furnace.

108,354, November 5, 1907, Canada. New and Useful Improvement in Casehardening Apparatus.

108,355, November 5, 1907, Canada. New and Useful Improvements in Casehardening Processes.

10,532, November 12, 1907, England. Improvements in Preparation of Carburizing Gas.

25,087, November 12, 1907, England. Improvements in Casehardening of Steel and Iron Articles.

236,007, March 19, 1908, Germany. Verfahren zur Herstellung eines zum Zementieren von Stahl-oder Eisenwaren dienenden, aus kohlenstoffhaltigen und aus indifferenten Gasen bestehenden Gasgemisches.

884,180, April 7, 1908, U. S. A. Improvements in Casehardening Apparatus. Applied for December 19, 1905, patented April 7, 1908. This covers improved gas carburizing machine with *exterior loading and discharge* of work and *apparatus for dilution of carburizing gas* by its absorption of diluent to prevent excessive carbon deposit. One diluent mentioned, for instance, ammonia.

884,181, April 7, 1908, U. S. A. Improvements in Casehardening. Applied for December 20, 1905, patented April 7, 1908. This is a companion patent to 884,180. It covers the *process* of diluting of carburizing gas by means of absorption of a diluent such as ammonia or other inert gas.

890,148, June 9, 1908, U. S. A. Improvement in Hardening Apparatus. This covers an automatic quenching apparatus.

389,547, July 2, 1908, France. Procédé et appareil de cementation.

961,305, June 14, 1910, U. S. A. Improvements in Annealing Apparatus. Applied for May 25, 1906, patented June 14, 1910. A companion patent to 1,065,697, a gas annealing process which may be carried out in the "Annealing Apparatus" which this patent covers. This is a rotating retort annealing machine, though not limited to rotation.

976,891, November 29, 1910, U. S. A. Improvements in Straightening and Hardening Apparatus for the straight hardening of shafting.

994,322, June 6, 1911, U. S. A. Improvements in Casehardening Apparatus. Applied for February 18, 1908, patented June 6, 1911. Covers a combination apparatus and process including a gas producer using charcoal or peat-moss, the resulting products of combustion, carbon monoxide, nitrogen, etc., being enriched by a hydrocarbon for carburizing, or, by by-passing the hydrocarbon enriching means, the producer gas of carbon monoxide, nitrogen and hydrogen may be simultaneously passed through another retort containing work to be annealed.

1,013,190, January 2, 1912, U. S. A. Casehardening Apparatus. Applied for May 17, 1906, patented January 2, 1912. A casehardening apparatus disclosing a process for gas carburizing using an open or crate-like rotating retort, entirely in the heating chamber of the furnace, having thick walls, i.e., in which the work in the open retort is directly in contact with the furnace atmosphere. The carburizing gas is introduced through the burner orifices (or it may be introduced otherwise) after the furnace walls and work-loaded retort have reached the carburizing heat and the air blast in the fuel lines has been shut off. The fuel gas, if suitable, may be the carburizing gas or a separate carburizing gas may be passed through the fuel burner orifices. Depth of car-

burizing may be governed by repeating the cycles of heating and carburization. The carburizing gas may or may not be diluted, say, for instance, by ammonia or other inert gas.

1,013,191, January 2, 1912, U. S. A. Casehardening Apparatus. Applied for May 25, 1906, patented January 2, 1912. This patent is similar to 1,013,190, but in this case no retort is used, the work to be carburized being stationary and being stacked directly in the heating chamber of the furnace. The carburizing gas being the same as the fuel gas; if suitable for carburizing, it may or may not be diluted by ammonia or other inert gas, or, if the fuel gas is not a suitable carburizing medium, ammonia or other diluent may be passed through a hydrocarbon-enriching fluid kept at constant level to obtain a uniform carburizing gas and to be independent of any possible unsuitable city or fuel gas as the carburizing medium.

1,033,815, July 30, 1912, U. S. A. Apparatus for Casehardening and other Metal Treatment. Applied for May 28, 1906, patented July 30, 1912. This covers a *machine* with rotating retort, wholly within the heating chamber. The retort is supported upon and driven by traction wheels which project into the heating chamber, but said traction wheels are mounted on driving shafts which are mounted exteriorly of the heating chamber. This machine, besides being intended for gas carburizing, is also employable for oxidizing steel, iron, copper and other materials by the use of suitable gas in each case.

1,045,161, November 26, 1912, U. S. A. Straightening Apparatus. A shaft straightening machine.

1,065,379, June 24, 1913, U. S. A. Improvements in treatment of steel, iron, etc. Applied for March 19, 1908, patented June 24, 1913. The *process* of heating cast iron, iron or steel of different grades in an atmosphere of ammonia to temperatures from about 900 to 1800 degrees Fahr. or below melting, to form a hard case or shell upon the metal articles. This is not limited to a rotating or a stationary retort, nor to any particular shape of retort.

This is the basic Nitriding Process patent. The practice of the process of nitriding has nowhere, even to this day, been changed in any essential from the disclosures of this patent.

1,065,697, June 24, 1913, U. S. A. Improvements in Annealing Processes, etc. Applied for May 25, 1906, divided and this application filed May 25, 1910, patented June 24, 1913. Annealing *Process* by treating in an inert gas, or in a carbonous gas if the articles had previously been subject to decarburization, or in a mixture of gases such as carbonous gas with ammoniacal constituents. For annealing steel or other metals. Covers rotating the articles, but not limited to rotation nor any particular shape of retort.

1,072,134, September 2, 1913, U. S. A. Continuous Casehardening Machine. Applied for August 17, 1908, patented September 2, 1913. An apparatus for *continuously casehardening, annealing or otherwise* heat treating metals and metal articles, comprising a furnace with revolving retort and magazine feed to feed the metal articles to the retort, in which they are acted upon by a gas or other fluid and then automatically discharged from the retort.

So far as is known, this patent discloses the first continuous gas carburizing machine.

1,092,925, April 14, 1914, U. S. A. Improvements in Hardening or Treatment of Steel, Iron, etc. Applied for July 12, 1907, patented April 14, 1914. The *process* of treating iron or steel by heating and subjecting to an atmosphere of ammonia and kerosene or other hydrocarbon and then to an atmosphere of ammonia alone. This in effect is a nitriding process superimposed upon a carburizing process, but the temperature is not changed and the two gas processes are separate steps, not simultaneous, thus differing from the later "Ni-Carb" process.

1,106,573, August 11, 1914, U. S. A. Improvements in Case Hardening Processes. Applied for May 25, 1906, divided, and this application filed December 19, 1911, patented August 11, 1914. This is a companion patent to 1,013,191, but this covers the *process* only as described in patent 1,013,191. The articles are heated directly by the heating gases (there being no retort), the heating is shut off, the work-containing heating chamber is filled with carburizing gas which carburizes the articles by means of their retained heat and the retained heat in the thick-walled oven. The other notes in connection with 1,013,191 apply in this case, such as the method of introducing the carburizing gas through the fuel burners, etc., the kind of carburizing gas and the provision for independence of possible unsuitable city or fuel gas as a carburizing medium.

1,152,959, September 7, 1915, U. S. A. Case Hardening Process. Applied for April 13, 1909, patented September 7, 1915. This is a *process* similar to that described in connection with the apparatus covered in patent 994,322. In this gas casehardening process, improvements in connection with a cheap carburizing gas produced in a gas producer are made. The gas is produced principally from air passing through incandescent charcoal mixed with peat moss. Independence from city or "line" gas is attainable.

1,337,244, April 20, 1920, U. S. A. Apparatus for Case Hardening. Applied for June 29, 1918, patented April 20, 1920. *Apparatus* for casehardening and other metal treatment. This covers the improved rotary gas carburizing machine with Nichrome retort, improvements to expedite charging and discharging. Not limited, however, to rotation.

1,471,618, October 23, 1923, U. S. A. Improvements in Furnaces. A furnace with revolving magazine for production heating of rods and shafts for forging, annealing, hardening, etc.

1,550,911, August 25, 1923, U. S. A. Gas Furnace with atmosphere control.

1,580,148, April 13, 1926, U. S. A. Barre-Preventing Means for Looms for textile industry.

1,629,581, May 24, 1927, U. S. A. Tools for Drilling Concrete.

291,300, December 3, 1927, England. Improvements in Apparatus for Singeing Cloth and Analogous Purposes.

1,655,233, January 3, 1928, U. S. A. Cloth-Singeing Apparatus.

1,713,543, May 21, 1929, U. S. A. Furnaces for Melting Metals. Cupola Type.

1,742,607, January 7, 1930, U. S. A. Gas Burners.

1,745,104, January 28, 1930, U. S. A. Carburizing Process. See Reissue 19,936.

- 493,474, March 8, 1930, Germany. Sengvorrichtung fuer Gewebe.
- 357,410, November 10, 1930, England. Improvements in Brass Melting Apparatus.
- 1,789,149, January 31, 1931, U. S. A. Brass Melting Furnace.
- 706,082, March 23, 1931, France. Appareil pour la fusion du Laiton (Brass Melter).
- 1,808,145, June 2, 1931, U. S. A. Brass Melting Apparatus.
- 318,996, January 19, 1932, Canada. Brass Melting Apparatus.
- 129,296, July 25, 1932, Austria. Drehrohr-Ofen zum schmelzen von Messing (Brass Melter).
- 403,792, October 31, 1932, England. Improvements in and relating to Furnaces, Bell Retort Type for carburizing, nitriding and other gas processes.
- 563,154, November 3, 1932, Germany. Schmelzofen fuer Messing (Brass Melter).
- 1,899,764, February 22, 1932, U. S. A. Gas Measuring Device. A direct reading Flow Meter.
- 1,921,128, August 8, 1933, U. S. A. Finish for Ferrous Articles. Applied for April 21, 1930, patented August 8, 1933. This covers a *new article of manufacture* of ferrous metal and the process. The article of whatever form, shape, size and analysis is provided with a nitrogenized foundation, furnace oxidized to a dark gray color and gun metalized, or nitrogenized and gun metalized without the intermediate furnace oxidation. The process is entirely a gas process.
- 1,938,599, December 12, 1933, U. S. A. Carburizing Apparatus. Applied for March 21, 1930, patented December 12, 1933. This is a companion patent to Reissue 19,936, but it covers *apparatus* for gas carburizing by mixtures of air and rich hydrocarbon such as propane.
- 1,941,807, January 2, 1934, U. S. A. Gas Bulb Heating Apparatus for heating electric lamp bulbs, etc.
- 1,995,314, March 26, 1935, U. S. A. Processes of casing Steel Articles. Applied for November 21, 1931, patented March 26, 1935. This covers the "*Ni-Carb*" process of gas casehardening by mixtures of ammonia gas and hydrocarbon gas. See also later patents on this process.
- 2,003,970, June 4, 1935, U. S. A. Bell Retort Furnaces. Applied for December 3, 1930, patented June 4, 1935. This covers the *Bell Retort Furnace*—for heat treating, annealing, carburizing, nitriding, etc., by mean of gas.
- 2,021,072, November 12, 1935, U. S. A. Continuous Heat Treating Machine. Applied for November 21, 1931, patented November 12, 1935. This covers a Continuous Heat Treating Machine—the reciprocating muffle type—for gas carburizing or other gas treating processes.
- 443,524, March 3, 1935, England. Improvements in and relating to the *Process of Casehardening Iron or Steel Articles*.
- Reissue 19,936, April 21, 1936, U. S. A. (Original 1,745,104.) Carburizing Process. Original applied for May 6, 1927, patented January 28, 1930. This Reissue applied for March 21, 1930, patented April 21, 1936. This covers the gas carburizing *process* using mixtures of air and propane or other rich hydrocarbon gases.

2,142,139, January 3, 1939, U. S. A. *Hardening Processes* for High Speed Steel Tools and Other Articles.

2,143,128, January 10, 1939, U. S. A. Needle Point Burners.

2,188,226, applied for February 24, 1938, patented January 23, 1940. This covers the "*Ni-Carb*" process in a more complete and advanced stage as compared to Patent No. 1,995,314.

One of his more recent developments is a Bell Oven. An Oven Furnace with a bell type muffle for general heat treating and such processes as carburizing, nitriding, "*Ni-Carbing*", clean hardening, bright annealing, brazing, etc.

He has also done much work in the direct reduction of iron ore without the addition of flux. New Jersey has a great deal of iron ore of high quality which he has hopes will some day be exploited in a truly economical manner.

President Gill then presented a certificate to Mr. Charles Strickler, who had been secretary of the York chapter for ten years and who was attending the convention this year as a guest of the ASM.

Mr. Gill then called on Mr. Charles Shapiro, chairman of the constitution and by-laws committee, to present to the membership the proposed changes in the constitution, as follows:

The proposed changes are shown in capitals.

Article I, Section 2 (a) now reads: A Founder Member shall be such person as the Board of Trustees shall determine has been instrumental in the founding of the Society or has rendered distinguished service to the Society.

This is to be changed to read as follows: A Founder Member shall be such person as the Board of Trustees shall determine has been instrumental in the founding of the Society AND has rendered distinguished service to the Society.

Article I, Section 2 (b) now reads: An Honorary Member shall be such person as the Board of Trustees shall determine is of acknowledged exceptional eminence. The total number of Honorary Members shall not exceed twenty-five (25) at any one time.

This is to be changed to read as follows: An Honorary Member shall be such person as the Board of Trustees shall determine HAS MADE EXCEPTIONAL CONTRIBUTIONS TO THE FIELD OF METALLURGY. The total number of LIVING Honorary Members shall not AT ANY TIME exceed twenty-five (25).

Article III, Section 2 now reads: The annual meeting of each local Chapter shall be held in the month of May of each year. The fiscal year of each local Chapter shall begin on the date of its annual meeting.

This is to be changed to read as follows: The annual meeting of each local Chapter shall be held in the month of APRIL OR May. The fiscal year of each local Chapter shall begin on the date of its annual meeting.

Article V, Section 7 now reads: Any member who shall have paid his dues for thirty-five (35) years, or who having paid his dues for twenty (20)

years shall have reached the age of seventy (70) years, shall thereafter be exempt from the payment of further dues.

This is to be changed to read as follows: Any member WHOSE DUES SHALL HAVE BEEN PAID FOR thirty-five (35) CONSECUTIVE years, or WHOSE DUES SHALL HAVE BEEN PAID FOR twenty (20) CONSECUTIVE years AND WHO SHALL HAVE REACHED THE AGE OF seventy (70) years, shall thereafter be exempt from the payment of dues. WITHIN THE MEANING OF THIS SECTION, REPRESENTATIVES OF SUSTAINING MEMBERS SHALL BE CONSIDERED AS MEMBERS.

Mr. Shapiro moved the adoption of the changes as proposed.

PRESIDENT GILL: These changes in the constitution are all for clarification, and if there is no objection the changes will stand as read.

ELECTION OF OFFICERS

J. P. GILL, *Presiding*: Complying with the Constitution, I appointed on March 11, 1940, the following Nominating Committee, selected from the list of candidates suggested by eligible chapter executive committees prior to March 1, 1940:

Charles H. Herty, Jr., Chairman (Lehigh Valley Chapter)
H. H. Lester (Boston Chapter)
Adam Steever (Chicago Chapter)
W. H. Swanger (Washington Chapter)
W. H. Phillips (Pittsburgh Chapter)
L. E. Raymond (New Haven Chapter)
C. B. Swander (St. Louis Chapter)

This committee met in New York City on May 22, 1940, and made the following nominations:

President—Oscar E. Harder, Battelle Memorial Institute, Columbus, Ohio.
(1 year)
Vice President—Bradley Stoughton, Lehigh University, Bethlehem, Pa.
(1 year)
Secretary—W. H. Eisenman, Cleveland.
(2 years)
Trustee—E. L. Bartholomew, United Shoe Machinery Corp., Beverly, Mass.
(2 years)
Trustee—C. Y. Clayton, Missouri School of Mines & Metallurgy, Rolla, Mo.
(2 years)

A report of these nominations duly appeared in THE REVIEW for May, 1940.

I have been informed by the Secretary that no additional nominations were received prior to July 15, 1940, for any of the vacancies

occurring on the Board of Trustees. Consequently the nominations were closed. I shall now call upon the Secretary to carry out the provision of the Constitution in respect to the election of National Officers.

W. H. EISENMAN: Conforming with the provisions and requirements of the constitution of the American Society for Metals, I hereby cast the unanimous vote of the members for the election of the aforementioned candidates, who were regularly nominated on May 22, 1940.

PRESIDENT GILL: The provisions of the Constitution having been complied with, I hereby declare the candidates heretofore named to be duly and unanimously elected to the several specified offices, the terms of each just elected beginning on the day following the close of this annual meeting.

President Gill then introduced the newly-elected officers.

Each year there are a few papers accepted for publication which, for one of several reasons, cannot be presented at the Convention. Mr. Eisenman then read these papers by title. They are

Alloys of Manganese and Copper: Electrical Resistance, by R. S. Dean and C. T. Anderson, Bureau of Mines, U. S. Dept. of Interior.

Alloys of Manganese and Copper: Hardening by Cold Work and Heat Treatment, by R. S. Dean and C. T. Anderson, Bureau of Mines, U. S. Dept. of Interior.

Alloys of Manganese-Copper and Nickel; Hardening in the Pseudo Binary System Cu-Mn-Ni, by R. S. Dean and C. T. Anderson, Bureau of Mines, U. S. Dept. of Interior.

The meeting was then adjourned and President Gill then introduced the chairman of the Campbell Memorial Lecture, Dr. Clyde E. Williams, director of the Battelle Memorial Institute, who in turn introduced Dr. Samuel L. Hoyt with the following remarks:

Mr. President, Members of the American Society for Metals and Friends—I am particularly honored in being privileged to address the fifteenth Campbell Memorial Lecture session, and I am particularly happy to have this opportunity because of my high regard for the lecturer as a friend and as a scientist. I think the lecturer represents the leadership among the men in the science of metals which all of us like to look up to, and we base our confidence in the future of the metal industry and the future of this Society.

As I look over the vast amount of talent here I have no question about the future of our industry and our ability in the future to get many more Campbell Memorial Lecturers of the type we have

had in the past. In fact the gathering together of a group of this type is a very significant thing.

When most of the rest of the world is engaged in the most destructive enterprise in history, it is really something to consider to realize that this group of outstanding people can gather together for their own education. The American Society for Metals has been an active force in education. Starting out twenty-some years ago as a small group of scientists, it has grown rapidly and education has become one of its outstanding activities. This education has been carried on through the publication of technical papers, through educational activities, and through lectures of the type we have today.

Edward DeMille Campbell was an outstanding teacher and scientist and so it is fitting that the memory of Edward DeMille Campbell should be linked with the American Society for Metals annually on an occasion of this sort.

The list of Campbell Memorial Lecturers is the list of outstanding people in the field of the science of metals. The present lecturer is highly fitted to become a member of that distinguished group. His experience as a teacher, as a research man, and as a commercial worker has fitted him into the position of leadership which all Campbell Lecturers have performed.

Through his lifetime of study and thought and observation, Sam Hoyt has developed a background to discuss this subject which he has chosen and which I think of as the philosophy of research. He terms it "*The scientific method in metallurgy.*"

I am therefore very happy to present Dr. Samuel L. Hoyt.

Dr. Hoyt then presented his lecture which is published in full in this issue of TRANSACTIONS beginning on page 757.

THE ANNUAL BANQUET

On Thursday evening, October 24, 1940, the American Society for Metals was host to 1275 members and guests at its annual banquet held in the ballroom of the Statler Hotel, Cleveland. Seated at the speakers' table were the following: L. F. Herron, metallurgist, James H. Herron Co.; H. J. French, in charge of alloy steel and iron development, International Nickel Co., and trustee of A.S.M.; J. H. Van Deventer, editor, *Iron Age*; F. B. Foley superintendent of research, Midvale Co., and trustee of A.S.M.; Marcus

Saxman, president, Latrobe Electric Steel Co.; S. L. Hoyt, member staff, Battelle Memorial Institute, Campbell Lecturer; K. R. Van Horn, research metallurgist, Aluminum Company of America, and treasurer of A.S.M.; F. C. Crawford, president, Thompson Products Co.; F. J. Griffiths, executive vice-president, Copperweld Steel Co.; W. P. Woodside, vice-president, Climax Molybdenum Co., and past president of A.S.M.; E. J. Kulas, president, Otis Steel Co.; R. J. Wysor, president, Republic Steel Corp.; C. F. Hood, president, American Steel & Wire Co.; J. P. Gill, chief metallurgist, Vanadium-Alloys Steel Co., and retiring president of the A.S.M.; T. M. Girdler, chairman of the board, Republic Steel Corp.; O. E. Harder, assistant director, Battelle Memorial Institute, and new president of the A.S.M.; E. E. Goodwillie, assistant to vice-president, Bethlehem Steel Co.; Wm. E. Umstattd, president, Timken Roller Bearing Co.; Roy C. McKenna, president, Vanadium-Alloys Steel Co.; W. Mathesius, vice-president in charge of operations, United States Steel Corp., and past trustee of the A.S.M.; Bradley Stoughton, professor of metallurgy, Lehigh University, and new vice-president of the A.S.M.; C. M. White, vice-president in charge of operations, Republic Steel Corp.; M. A. Grossmann, director of research, Carnegie-Illinois Steel Corp., and trustee of the A.S.M.; D. S. Clark, instructor of mechanical engineering, California Institute of Technology, and trustee of the A.S.M.; E. L. Bartholomew, metallurgist, United Shoe Machinery Corp., and new trustee of the A.S.M.; W. H. Eisenman, secretary of the American Society for Metals.

President Gill acted as toastmaster and presented W. P. Woodside with the Past President's medal and a certificate of appreciation to Dr. S. L. Hoyt for the Campbell Memorial Lecture.

President Gill introduced T. M. Girdler, Chairman of the Board, Republic Steel Corp., who introduced the principal speaker of the evening, General Hugh Johnson.



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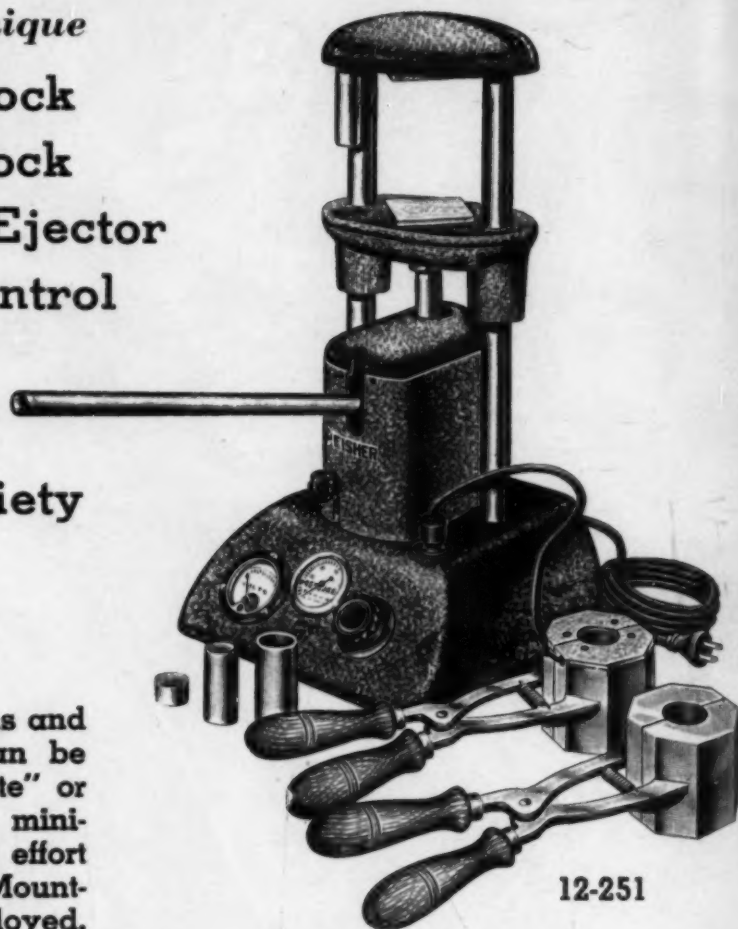
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1.—That the names and addresses of the publisher, editor, managing editor, and business managers are: Publisher, American Society for Metals, 7301 Euclid Ave., Cleveland, O.; Editor and Managing Editor, Ray T. Bayless, 7301 Euclid Ave., Cleveland, O.; Business Manager, W. H. Eisenman, 7301 Euclid Ave., Cleveland, O.

2.—That the owner is: The American Society for Metals, 7301 Euclid Ave., Cleveland, Ohio, which is an educational institution, the officers being: President, J. P. Gill; Vice-President, O. E. Harder; Secretary, W. H. Eisenman; Treasurer, K. R. Van Horn; Trustees: W. P. Woodside, F. B. Foley, D. S. Clark, H. J. French, M. A. Grossmann. All officers as above, 7301 Euclid Ave., Cleveland, Ohio.

3.—That the known bondholders, mortgagees, and other security holders owning or holding 1 per cent or more of total amount of bonds, mortgages, or other securities are: none.

4.—That the two paragraphs next above, giving the names of the owners, stockholders, and security holders, if any, contain not only the list of stockholders and security holders as they appear upon the books of the company but also, in cases where the stockholder or security holder appears upon the books of the company as trustee or in any other fiduciary relation, the name of the person or corporation for whom such trustee is acting, is given; also that the said two paragraphs contain statements embracing affiant's full knowledge and belief as to the circumstances and conditions under which stockholders and security holders who do not appear upon the books of the company as trustees, hold stock and securities in a capacity other than that of a bona fide owner; and this affiant has no reason to believe that any other person, association, or corporation has any interest direct or indirect in the said stock, bonds, or other securities than as so stated by him. Ray T. Bayless, managing editor, sworn to and subscribed before me this 1st day of October, 1940.

(Seal) Arthur T. Wehrle, notary public.

(My commission expires January 20, 1941.)

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